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*TRANSACTIONS*
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# IMPACT PROPERTIES OF SOME LOW ALLOY NICKEL STEELS AT TEMPERATURES DOWN TO —200 DEGREES FAHR.

By T. N. Armstrong and A. P. Gagnebin

*Abstract*

*For applications involving temperatures down to 75 degrees below zero Fahr., there are several steels, in both the wrought and cast form, that resist low temperature embrittlement sufficiently to perform satisfactorily. For lower temperatures, austenitic stainless steels and non-ferrous metals have been employed, but with the increased use of low temperature processes, involving temperatures of 100 degrees below zero Fahr. and below, there has been a search for less expensive materials.*

*Data are presented showing that several normalized and drawn nickel and nickel-molybdenum steels, produced both experimentally and commercially, possess good resistance to impact at these lower temperatures. Specimens machined from rolled bars, from cast coupons and from deposited weld metal were tested in impact, Charpy specimens with keyhole notch, at temperatures as low as 200 degrees below zero Fahr.*

THE widest use of steels for low temperature applications has been for operating temperatures between —40 and —100 degrees Fahr. (—40 and —75 degrees Cent.). There has been a recent trend to utilize even lower temperatures for fractional distillation of liquids, and specifications are now common which require steels that retain a fair degree of shock resistance at temperatures

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, T. N. Armstrong is a member of the staff of the development and research division and A. P. Gagnebin is member of the staff of the research laboratory, International Nickel Co., New York. Manuscript received June 26, 1939.

down to —150 degrees Fahr. (—100 degrees Cent.). When such low temperatures were first utilized, engineers turned to austenitic chromium-nickel steels and nonferrous metals, such as monel and pure nickel, as these materials are only slightly embrittled at extremely low temperatures. With increase in use of temperatures below —100 degrees Fahr. (—75 degrees Cent.), there has been a search for materials that may be substituted for these high alloys in the interest of economy. In this paper data are presented that show the suitability of certain of the low alloy nickel alloy steels for service at temperatures as low as —200 degrees Fahr. (—130 degrees Cent.).

## Method of Test

Unless otherwise indicated the test methods used in this investigation were as follows:

The Charpy machine was equipped with a fixture to position the specimens properly so that only about two seconds elapsed from the time a specimen was removed from the refrigerating bath until it was struck by the pendulum. All tests were run using the full energy of the machine, 240 foot-pounds.

Temperatures down to and including —100 degrees Fahr. (—75 degrees Cent.) were obtained by adding dry ice (solid carbon dioxide) to alcohol contained in an insulated beaker. Lower temperatures were obtained by chilling petroleum ether with liquid air as shown in Fig. 1. The beaker containing the petroleum ether was partly immersed in the liquid air until the desired temperature was reached and the temperature then maintained by adjusting the height of the beaker above the surface of the liquid air. The petroleum ether was stirred gently to assure a uniform temperature. In using petroleum ether special precautions had to be taken as the volatile products in certain concentrations are explosive.

Impact specimens with keyhole notch conformed in detail to Fig. 2, page 673 of the 1939 A.S.M. Metals Handbook. Vee notched specimens conformed to details given in Fig. 10, page 135 of the 1939 A.S.M. Metals Handbook.

Values given in the paper are the averages of two or more tests and when there was considerable scatter in results at a specific temperature, all values are shown.

In testing steels in impact at subzero temperatures, resistance to impact generally decreases gradually with decrease in tempera-

ture until some temperature is reached where both brittle and tough fractures, resulting in both high and low impact values, will occur in the same steel. Subsequent tests at lower temperatures will give values that generally do not exceed the lowest values secured at the temperature producing the spread in impact results. Curves indicat-

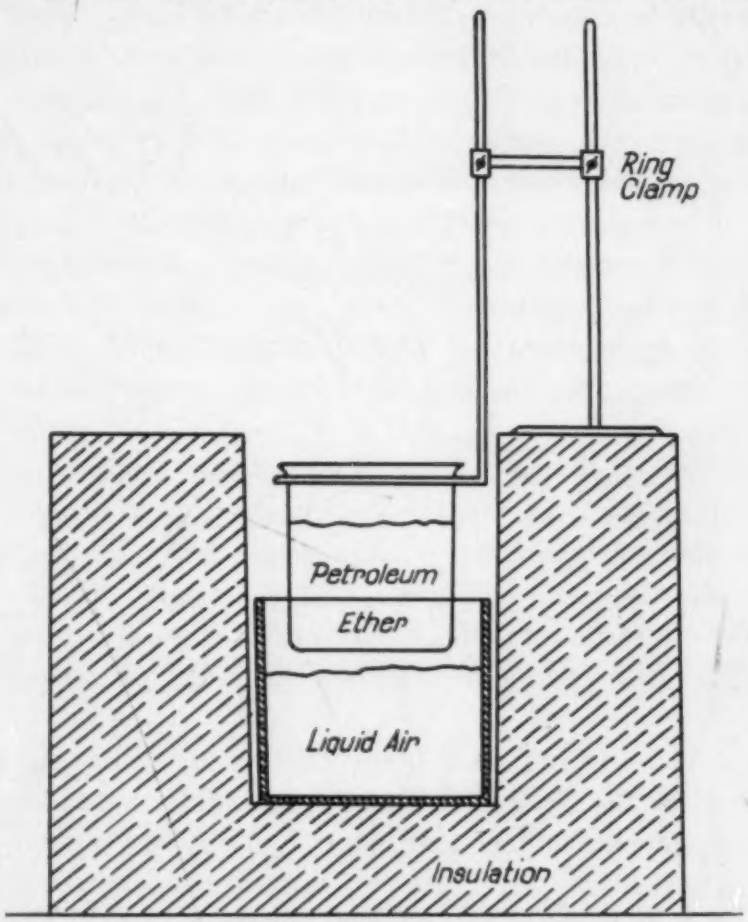


Fig. 1—Sketch of Apparatus for Chilling Specimens From Temperatures of —100 Degrees Fahr. to —200 Degrees Fahr.

ing individual values will show one or more splits or discontinuities at these points. That this phenomenon is not unusual was first pointed out by Kinzel (1)[1] who reported results from Izod tests. Less sensitive tests, such as Charpy with keyhole notched specimens, show the same discontinuities but not necessarily at the same temperatures. It is probable that all ferritic steels will show such behavior providing temperatures are carried sufficiently low and also providing the testing temperatures coincide with the temperatures at which scatter occurs.

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

## Factors Affecting Low Temperature Impact Properties

*Carbon*—Results from some of the early work in this field led to the belief that carbon was one of the controlling influences, and for steels that were to be treated by normalizing and drawing, best results would be secured by limiting carbon to low values (2), (3). More recent work has indicated that resistance to embrittlement at low temperatures depends not so much upon the quantity of carbon present as upon the degree of dispersion. Hence, steels which have a high degree of carbide refinement, such as is obtained by quenching and drawing, can tolerate a much higher carbon content than steels that are treated by normalizing or by annealing. Also, this infers that steels containing very low carbon will not be susceptible to improvement by heat treatment to the same degree as steels containing appreciable quantities of carbon. Results of tests indicate that this is true.

*Deoxidation*—From the results of work by Herty (4) and others, it is apparent that degree of deoxidation has a decided influence upon low temperature impact properties. Silicon-killed steels give better results than rimmed steels, and thoroughly killed steels containing definite amounts of aluminum in solution appear to retain toughness to lower temperatures than steels deoxidized in any other manner.

*Grain Size*—It has been observed that steels having fine grain, as measured by the McQuaid-Ehn test, generally have decidedly better low temperature impact resistance than coarse-grained steels (4), and the influence of grain size on impact properties appear to be much greater at subzero temperatures than at normal temperatures. Occasionally, however, a steel having a fairly coarse grain will possess good resistance to low temperature impact. In this connection, it is believed pertinent to call attention to the fact that a steel having coarse McQuaid-Ehn grain size does not necessarily have unrefined carbides and, also, steels having a fine McQuaid-Ehn grain size have usually been treated with sufficient aluminum to retain definite quantities of aluminum in solution. This indicates that carbide dispersion and aluminum deoxidation may overshadow grain size *per se*.

*Nickel*—One of the most effective means of improving the low temperature impact properties of steel is by the addition of nickel (5). This influence is exerted alike on cast and wrought steels of any carbon content and in any condition of heat treatment. In the

past, nickel alloy steels have been used extensively for temperatures down to —75 degrees Fahr. (—60 degrees Cent.) and data reported herein indicate that the field of usefulness of certain low alloy nickel steels may be extended to —200 degrees Fahr.

*Aging*—In some of the early work on low temperature performed in the research laboratory of the International Nickel Company, it was found that different lots of commercial steels having almost identical analyses would give widely scattered results when tested in impact at —50 degrees Fahr. (—45 degrees Cent.). As the low temperature toughness of some of these steels, air-cooled from the draw, exceeded that of those water-quenched from the draw, indications were that the form and distribution of the soluble constituents in ferrite had an effect. This indicated the existence of some critical cooling rate to secure greatest toughness. A 0.19 per cent carbon, 2 per cent nickel steel showing such behavior was normalized at 1650 degrees Fahr. (900 degrees Cent.) and given a solution treatment consisting of heating three hours at 1200 degrees Fahr. (650 degrees Cent.) and water quenching. Specimens were then reheated at lower temperatures for different lengths of time. The results of this experiment were as follows:

| Reheating Temperature | Charpy Impact at —50 Degrees Fahr. (ft. lbs.) after Reheating. Steel first Normalized at 1650 Degrees Fahr., Drawn 1200 Degrees Fahr., and Water-quenched. | | |
|---|---|---|---|
| | 1 hour | 2 hours | 7½ hours |
| No reheating | 5, 6 | | |
| 200 | 10 | 6 | 4 |
| 300 | 5 | 7 | 4 |
| 400 | 21 | 26 | 11 |
| 500 | 13 | 7 | 6 |
| 600 | 4 | 6 | 4 |
| 800 | 4 | 3 | 3 |
| 1000 | 5 | 6 | 5 |

With this steel the critical aging treatment was achieved by reheating at 400 degrees Fahr. (200 degrees Cent.) for 1 or 2 hours, which substantially raised the Charpy values at —50 degrees Fahr. (—45 degrees Cent.). This same reheating temperature was used on several other steels without success which indicates that apparently this phenomenon does not take place at the same temperature in all steels. The results did suggest a relation between low temperature brittleness and aging phenomenon.

It is considered significant that low temperature toughness

could be influenced by reheating temperatures as low as 400 degrees Fahr. (200 degrees Cent.), which precludes the movement of any but the so-called mobile elements such as carbon, oxygen, nitrogen and hydrogen. Pursuing the idea of controlling the disposition of

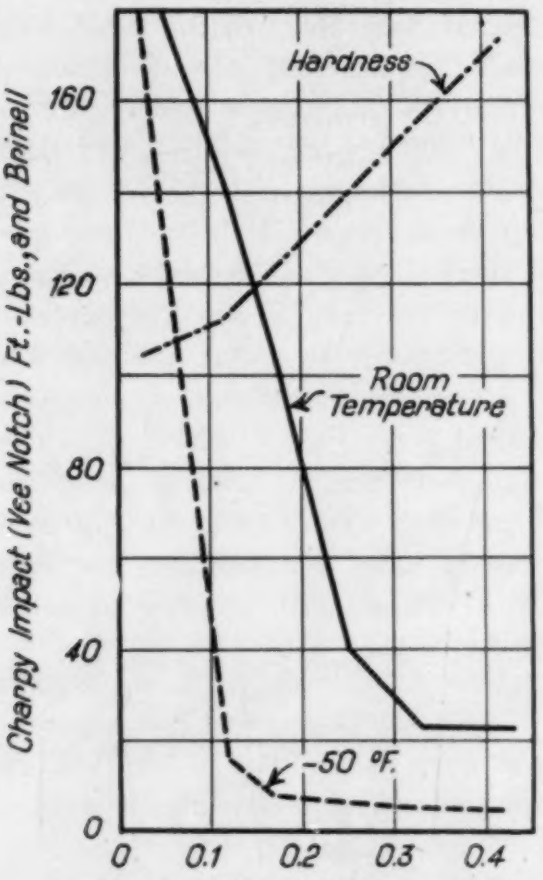


Fig. 2—Effect of Carbon on Hardness and Impact Properties of Carbon Steels.

the mobile elements by melt control, two induction furnace melts were made, one deoxidized with manganese and silicon and the other deoxidized with manganese and silicon followed by aluminum. Results are shown in the following tabulation:

| Per Cent | | | | | Charpy Impact, Ft. Lbs., (Vee Notch) | | | |
|---|---|---|---|---|---|---|---|---|
| | | | | | As Forged | | Normalized 1 hr. 1650° F. | |
| C | Ni | Mn | Si | Al added | Rm. Temp. | —50° F. | Rm. Temp. | —50° F. |
| 0.20 | 2.01 | 0.65 | 0.12 | .... | 79 | 26½ | 80½ | 22½ |
| 0.19 | 2.14 | 0.66 | 0.12 | 0.04 | 75½ | 62½ | 75½ | 56 |

These results indicate the importance of the degree of deoxidation to secure high impact resistance at subzero temperatures.

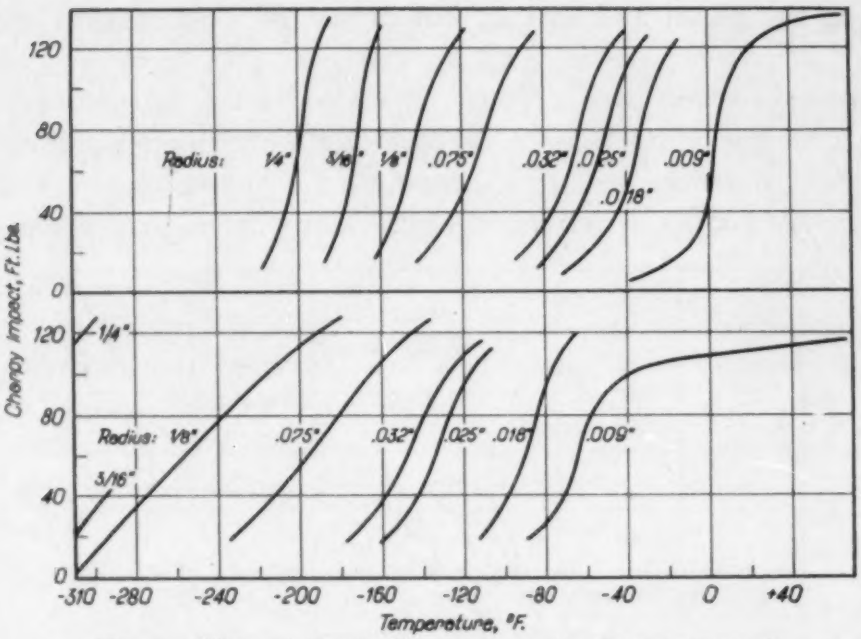


Fig. 3—Effect of Notch Sharpness on the Impact Properties of a Carbon and a Nickel Steel.

*Strength with Toughness*—To determine the effect of carbon in normalized steels, a series of induction furnace melts were made with carbon ranging from practically zero to 0.40 per cent. The Charpy values at room temperature and at —50 degrees Fahr. (—45 degrees Cent.) are shown in Fig. 2. These results indicate that even small increases in carbon, other things being equal, cause a rapid decrease in impact resistance at room temperature and an even more severe decrease in impact resistance at —50 degrees Fahr. (—45 degrees Cent.). To determine the effect of nickel when substituted for carbon to obtain a given hardness or strength, the carbon and nickel contents were adjusted in a series of five melts to give a Brinell hardness of 150 after normalizing. The results at room temperature and at —50 degrees Fahr. (—45 degrees Cent.) are shown as follows:

| BHN | Ni | C | Charpy, ft. lbs., (Vee notch) Rm. Temp. | —50° F. |
|---|---|---|---|---|
| 150 | 0 | 0.33 | 25 | 5 |
| 150 | 1 | 0.22 | 85 | 20 |
| 150 | 2 | 0.16 | 95 | 43 |
| 150 | 3 | 0.12 | 110 | 70 |
| 150 | 4 | 0.04 | 172 | 170 |

From these results it is evident that both good tensile strength and resistance to low temperature impact can be secured in steel by

balancing the nickel and carbon, but cannot be secured with carbon alone.

*Stress Concentration (Notch Effect)*—Successful performance of steels at low temperatures depends not only upon composition, deoxidation practice, heat treatment and the operating temperature but upon distribution of stress as well. Many steels that would per-

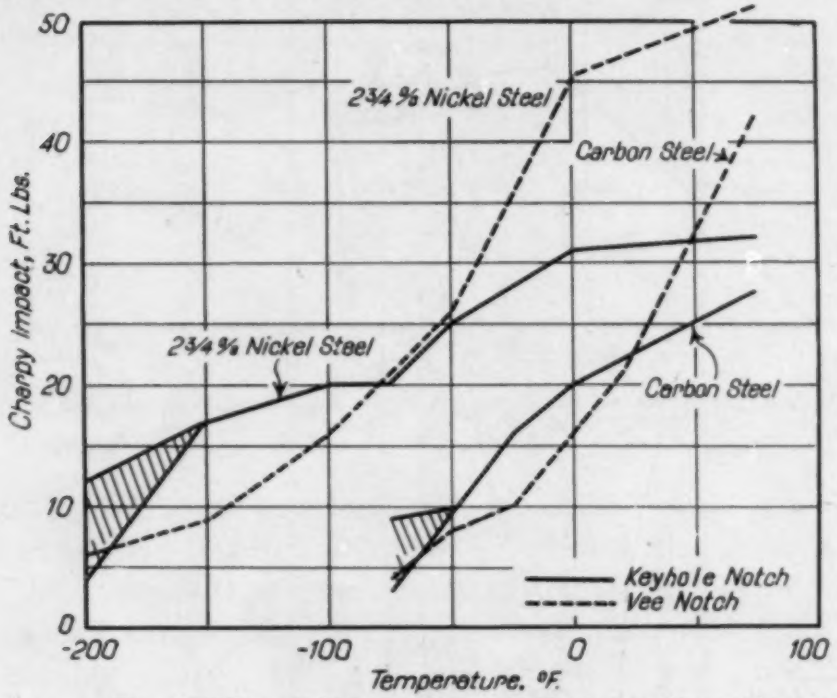


Fig. 4—Effect of Different Types of Notches on Impact Properties of Two Normalized and Drawn Cast Steels Containing 0.18 Per Cent Carbon.

form satisfactorily at low temperatures providing there was freedom from stress concentration cannot be used as they show appreciable embrittlement when tested in impact with notched bar. The effect at —50 degrees Fahr. (—45 degrees Cent.) of increasing the sharpness of notch in specimens of a normalized 0.15 per cent carbon steel and a normalized 0.15 per cent carbon, 2 per cent nickel steel is shown in Fig. 3. Most commercial specifications are based on values secured from testing Charpy bars with keyhole notch. That this type of notch is not so sensitive to differences in either composition or temperature as the Vee notch, as indicated by the curves shown in Fig. 4, but so long as the keyhole notch is specified by fabricators and users of chilling equipment, it is logical to use this type specimen in studying the low temperature impact properties of steels.

*Steels Suitable for Service at —75 Degrees Fahr.*—Properties of both cast and rolled nickel alloy steels have been reported by several investigators and have been assembled in a comprehensive form by French and Sands (6). These data indicate that toughness sufficient to meet most requirements at —75 degrees Fahr. (—60 degrees Cent.) can be secured in fine-grained plate steels containing not over 0.20 per cent carbon and a minimum of 2 per cent nickel, in the as-rolled condition, although higher values result from normalizing and drawing. Results are also given showing that cast steels containing a maximum of 0.25 per cent carbon and not less than 2.25 per cent nickel will consistently give values in excess of 15 foot-pounds Charpy, keyhole notch, at —75 degrees Fahr. (—60 degrees Cent.), after normalizing and drawing.

In studying the properties of normalized steels at —150 degrees Fahr. (—100 degrees Cent.), it was found that the steels widely used for —75 degrees Fahr. (—60 degrees Cent.) service (that is, 0.20 per cent carbon, 2 per cent nickel rolled steel and 0.20 per cent carbon, 2.50 per cent nickel cast steel) lose a considerable portion of their resistance to impact at temperatures below —100 degrees Fahr. (—75 degrees Cent.). To determine the optimum carbon and nickel contents and also to determine which of the commercially available nickel alloy steels were satisfactory for this service, a number of induction furnace melts were made and tested in impact at temperatures down to —200 degrees Fahr. (—130 degrees Cent.). Several commercial steels, both wrought and cast, were included in the test.

## Cast Steels

A series of cast steels containing from 0 to 5.50 per cent nickel and carbon within the range of 0.10 to 0.15 per cent were melted in the induction furnace and 0.08 per cent aluminum added to the ladle after furnace deoxidation with manganese and silicon. The metal was poured into keel type test blocks, the bottom sections of which were approximately 1x1x8 inches and comprised the test coupons. These coupons were double normalized and drawn at 1200 degrees Fahr. (650 degrees Cent.). Two experimental heats containing both nickel and molybdenum and two commercial steels, one containing 3.50 per cent nickel and the other containing nickel, molybdenum and vanadium, were included in the same series of

**Table I**
**Cast Steels, Composition and Room Temperature Tensile Properties**
(Double Normalized and Drawn, 1200 Degrees Fahr.)

| | Per Cent | | | | | | Yield Point p.s.i. | Tensile Strength p.s.i. | Per Cent Elon. | Per Cent Reduction of Area |
|---|---|---|---|---|---|---|---|---|---|---|
| | C | Mn | Si | Al Added | Ni | Mo | | | | |
| A | 0.18 | 0.67 | 0.40 | 0.05 | nil | nil | 41,200 | 62,000 | 32.0 | 45.0 |
| B | 0.14 | 0.70 | 0.38 | 0.05 | nil | nil | 39,500 | 64,200 | 36.0 | 59.0 |
| C | 0.13 | 0.65 | 0.35 | 0.08 | nil | nil | 38,600 | 54,200 | 42.0 | 65.5 |
| D | 0.12 | 0.65 | 0.35 | 0.08 | 1.00 | nil | 41,500 | 59,500 | 34.5 | 63.0 |
| E | 0.14 | 0.65 | 0.35 | 0.08 | 2.00 | nil | 50,200 | 69,000 | 37.5 | 67.5 |
| F | 0.15 | 0.58 | 0.30 | 0.016 | 2.90 | nil | 61,200 | 76,800 | 33.0 | 55.5 |
| G | 0.11 | 0.64 | 0.34 | 0.08 | 4.95 | nil | 48,200 | [illegible] | 26.0 | 45.5 |
| H | 0.11 | 0.65 | 0.36 | 0.08 | 3.45 | 0.27 | 66,200 | 87,200 | 26.5 | 59.0 |
| K | 0.03 | 0.22 | 0.04 | 0.01 | 4.36 | 0.45 | 66,250 | 83,250 | 28.[illegible] | 71.9 |
| L | 0.13 | 0.65 | 0.35 | 0.08 | 3.50 | nil | 53,500 | 79,500 | 32.0 | 60.0 |
| M | 0.10 | 0.65 | 0.35 | 0.08 | 4.06 | nil | 48,000 | 71,500 | 35.0 | 65.6 |
| N | 0.11 | 0.65 | 0.35 | 0.08 | 2.82 | 0.31 | 59,700 | 71,300 | 34.5 | 69.5 |

tests. Room temperature tensile properties were determined for all steels and Charpy impact properties, using both keyhole and Vee notched specimens, were determined for temperatures down to —200 degrees Fahr. (—130 degrees Cent.). Results of the tests are given in Tables I and II.

All of the carbon steels were embrittled at temperatures of —75 degrees Fahr. (—60 degrees Cent.) and only one of the three showed any indication of appreciable resistance to impact at —50 degrees Fahr. (—45 degrees Cent.), based upon the assumption that —75 degrees Fahr. (—60 degrees Cent.) was the embrittling temperature as no tests were made on this steel at —50 degrees Fahr. (—45 degrees Cent.). In addition to low impact resistance, these steels all possess low strength. This deficiency in both strength and resistance to embrittlement indicates the unsuitability of cast carbon steel for subzero temperature service.

**Table II**
**Cast Steels—Low Temperature Charpy Impact Properties**
(Double Normalized and Drawn, 1200 Degrees Fahr.)

| | Keyhole Notch | | | | | | Vee Notch | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|
| | 70° F. | —50° | —75° | —100° | —150° | —200° | 70° F. | —50° | —100° | —150° | —200° |
| A | 22 | 6 | 6 | .. | .. | .. | 30 | 4 | .. | .. | .. |
| B | 22 | 9 | 3 | .. | .. | .. | 38 | 7 | .. | .. | .. |
| C | 44/47 | .. | 19/3 | .. | .. | .. | 71 | 9 | 2 | 1 | .. |
| D | 39/41 | .. | .. | 16/21 | 2/3 | .. | 73 | 23 | 8 | 2 | 1 |
| E | 37/41 | .. | .. | 19/18 | 16/15 | 7½/2½ | 68 | 57 | 13 | 5 | 4 |
| F | 30/32 | 25/25 | 21/19 | 21/19 | 17/17 | 12/4 | 52 | 26 | 15 | 8 | 6 |
| G | 30/35 | .. | .. | 20/18 | 19/15 | 10/8½ | 54 | 54 | 15 | 6 | 4 |
| H | 33/36 | .. | .. | 26/22 | 18/14 | 11/10 | 53 | 38 | 18 | 11 | 5 |
| K | 48 | .. | .. | 46/38 | 31/33 | 4/6 | 94 | 64 | 45 | 9 | 4 |
| L | 39/33 | .. | .. | 23/24 | 20/15 | 20/17 | 74 | 37 | 27 | 15 | 10 |
| M | 36/32 | .. | .. | 26/26 | 18/16 | 19/16½ | 65 | 34 | 22 | 13 | 6 |
| N | 43/46 | .. | .. | 31/33 | 28/27 | 20/19 | 71 | 51 | 30 | 18 | 11 |

Results in Table II indicate that the temperature at which steel is embrittled decreases as the nickel content is increased up to approximately 4 per cent nickel. A combination of nickel and molybdenum appears to be very effective in promoting shock resistance at low temperatures providing the percentages of carbon and alloy are not too high, resulting in sacrifice of toughness for hardness. This appears to be the case with the 3.50 per cent nickel-molybdenum steel, the 5 per cent nickel steel and the nickel-molybdenum-vanadium steel, as these steels are not as tough at —200 degrees Fahr. (—130 degrees Cent.) as steels not quite so heavily alloyed, although all the steels containing 2 per cent or more nickel have equal or better than 15 foot-pounds Charpy at —150 degrees Fahr. (—100 degrees Cent.) when tested with keyhole notch.

Optimum low temperature impact properties were secured with heats L, M, and N, which leads to the conclusion that double normalized and drawn cast steels containing from 3.50 to 4 per cent nickel and 2.75 per cent nickel and 0.30 per cent molybdenum retain an appreciable degree of resistance to embrittlement at temperatures down to —200 degrees Fahr. (—130 degrees Cent.).

In order to make a comparison of normalized and drawn, and quenched and drawn cast steels at low temperatures and also to determine if the effect of treatment was the same as for wrought steels, four cast steels containing from 0.12 to 0.14 per cent carbon and 0, 1, 2 and 3.50 per cent nickel respectively were given a 1750 degrees Fahr. (950 degrees Cent.) normalizing treatment, followed by reheating to just above the critical, quenching in oil and drawing at 1200 degrees Fahr. (650 degrees Cent.). Composition and properties are given in Table III. It is most surprising to observe that impact values do not equal those of normalized and drawn steels shown in Table II with the single exception of the carbon steel tested with Vee notch. The only explanation that can be offered is that carbon is so low in these steels that its effect is masked by other factors. Certainly the indications are that there is little value in quenching cast steels containing less than 0.15 per cent carbon for low temperature service unless it is desired to secure higher tensile strength and yield point.

Aluminum deoxidation is generally accepted as improving the low temperature properties of wrought steels. It might be assumed that cast steels would be affected in like manner; however, results published by Simms (7) and by Gagnebin (8) show that there are

**Table III**
**Low Temperature Impact Properties of Cast Nickel Steel**
Normalized 1750° F., Oil-Quenched and Drawn at 1200° F.

| Per Cent C | Per Cent Ni | Quenched from ° F. | Yield Point p.s.i. | Tensile Strength p.s.i. | Per Cent Elong. | Per Cent R. A. | Charpy Impact, ft. lbs. — Keyhole Notch: 70° F. | −75° F. | −100° F. | −150° F. | −200° F. | Vee Notch: 70° F. | −50° F. | −75° F. | −100° F. | −150° F. | −200° F. |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 0.13 | .... | 1650 | 40,800 | 57,000 | 41.2 | 68.0 | 45, 33 | 6, 5 | 15, 3 | 1, 2 | .. | 75, 79 | 26, 21 | 16 | 8 | 2 | .. |
| 0.12 | 1.03 | 1600 | 46,000 | 63,200 | 40.0 | 72.0 | 42, 41 | .. | 21, 26 | 24, 17 | 4, 1¼ | 68 | 48 | .. | 21 | 4 | 3 |
| 0.14 | 2.02 | 1550 | 57,000 | 73,500 | 34.5 | 63.0 | 34, 38 | .. | 34, 24 | 24, 24 | 14, 15 | 67 | 35 | .. | 23 | 11 | 8 |
| 0.13 | 3.50 | 1525 | 67,500 | 93,500 | 30.0 | 53.5 | 36, 33 | .. | 23, 24 | 19, 21 | 17, 11 | 55 | 35 | .. | 21 | 15 | 6 |

**Table V**
**Composition and Tensile Properties of Experimental Wrought Steels**

| No. | Composition, Per Cent — C | Mn | Si | Ni | Mo | Al* | Heat Treatment Degrees F. | Mechanical Properties — Yield Point, psi., 0.01" extension | Tensile Str. p.s.i. | % Elon. in 2" | % Red. Area | Bhn | McQ-Ehn | Charpy Impact, Ft. Lbs., Keyhole Notch — Room Temp. | 0° F. | −50° F. | −100° F. | −150° F. | −200° F. |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| AA | 0.09 | 0.57 | (0.15) | 0 | .... | 0 | 1675 AC, 1200 AC | 36,600 | 52,000 | 45.0 | 75.0 | 95 | 4 | 61, 62 | 19, 22, 46, 47 | 5, 6 | 2, 3 | .... | .... |
| AA1 | 0.09 | 0.57 | (0.15) | 0 | .... | 0.08 | 1675 AC, 1200 AC | 38,500 | 51,500 | 46.0 | 77.6 | 95 | 7 | 65, 65 | 23, 64, 65 | 7, 6 | 3, 3 | .... | .... |
| AB | 0.09 | (0.60) | (0.15) | 1.97 | .... | 0 | 1600 AC, 1200 AC | 46,400 | 61,600 | 40.5 | 74.3 | 117 | 3–4 | 57, 61 | .... | 40, 45 | 35, 36 | 3, 11 | .... |
| AB1 | 0.09 | (0.60) | (0.15) | 1.97 | .... | 0.08 | 1600 AC, 1200 AC | 48,100 | 62,700 | 41.5 | 76.8 | 120 | 7 | 63, 65 | .... | 55, 58 | 46, 48 | 33, 39 | 3, 29 |
| AC | 0.09 | 0.62 | (0.15) | 3.57 | .... | 0 | 1550 AC, 1200 AC | 50,000 | 68,800 | 38.5 | 70.7 | 133 | 4–5 | 57, 58 | .... | 43, 44 | 36, 41 | 32, 34 | 22, 28 |
| AC1 | 0.09 | 0.62 | (0.15) | 3.57 | .... | 0.08 | 1550 AC, 1200 AC | 53,100 | 69,200 | 38.0 | 75.0 | 133 | 7 | 59, 62 | .... | 49, 59 | 45, 48 | 41, 42 | 35, 36 |
| AD | 0.16 | (0.60) | 0.15 | 0 | .... | 0 | 1650 AC, 1200 AC | 37,700 | 57,400 | 42.0 | 70.7 | 106 | 2 | 52, 53 | 11, 10, 36, 36 | 4, 4 | 2, 2 | .... | .... |
| AD1 | 0.16 | (0.60) | 0.15 | 0 | .... | 0.08 | 1650 AC, 1200 AC | 39,100 | 56,600 | 42.0 | 70.9 | 103 | 7 | 56, 57 | 43, 44 | 7, 6, 6, 37 | 3, 3 | .... | .... |
| AE | 0.16 | (0.60) | (0.15) | 1.99 | .... | 0 | 1575 AC, 1200 AC | 46,600 | 66,400 | 36.5 | 69.7 | 128 | 3–4 | 48, 52 | .... | 31, 34 | 27, 31 | 3, 3 | .... |
| AE1 | 0.16 | (0.60) | (0.15) | 1.99 | .... | 0.08 | 1575 AC, 1200 AC | 50,500 | 66,900 | 39.5 | 72.5 | 130 | 7 | 55, 55½ | .... | 41, 45 | 35, 36 | 29, 29 | 6, 23 |
| AF | 0.17 | (0.60) | (0.15) | 3.61 | .... | 0 | 1525 AC, 1200 AC | 47,500 | 72,400 | 36.0 | 66.6 | 140 | 3–4 | 47, 48 | .... | 32, 36 | 26, 27 | 25, 26 | 15, 18 |
| AF1 | 0.17 | (0.60) | (0.15) | 3.61 | .... | 0.08 | 1525 AC, 1200 AC | 50,500 | 74,000 | 36.5 | 70.3 | 142 | 7–8 | 50, 52 | .... | 43, 43 | 37, 40 | 29, 31 | 26, 30 |
| AG | 0.17 | (0.60) | (0.15) | 2.83 | 0.31 | 0 | 1525 AC, 1200 AC | 64,600 | 83,600 | 28.0 | 68.2 | 176 | 3–4 | 47, 47½ | .... | 35, 35 | 31, 32 | 24, 26 | 2, 3 |
| AG1 | 0.17 | (0.60) | (0.15) | 2.83 | 0.31 | 0.08 | 1525 AC, 1200 AC | 67,500 | 81,500 | 31.5 | 71.1 | 168 | 7–8 | 50, 50½ | .... | 41, 43 | 37, 41 | 30, 32 | 29, 30 |

*Percentage of Aluminum added to ladle.
Numbers in parentheses are estimated values.

**Table IV**
**Effect of Aluminum Deoxidation on Properties of Two Cast Nickel Steels**

| | Composition Per Cent | | | | | Tensile Properties | | | | Charpy Impace, Ft. lbs. Keyhole Notch | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | C | Mn | Si | Ni | Al added | Yield Point p.s.i. | Tensile Strength p.s.i. | Per Cent Elong. | Per Cent Red. Area | Room Temp. | —50° F. | —75° F. |
| 1 | 0.24 | 0.62 | 0.42 | 2.39 | nil | 61,500 | 81,700 | 32.0 | 53.5 | 29 | 23 | 19 |
| 1A | 0.24 | 0.62 | 0.42 | 2.39 | 0.065 | 64,500 | 82,500 | 29.0 | 45.0 | 21 | 15 | 14 |
| 2 | 0.30 | 0.68 | 0.38 | 3.14 | nil | 64,000 | 85,000 | 31.0 | 58.5 | 33 | 22 | 20 |
| 2A | 0.30 | 0.68 | 0.38 | 3.14 | 0.065 | 65,300 | 85,000 | 28.5 | 47.5 | 26 | 15 | 12 |

certain hazards involved in adding aluminum to cast steel. Such additions may lead to embrittlement resulting from formation of a grain boundary network of sulphides, and this structure is not only deleterious to impact resistance but also impairs ductility.

Results shown in Table IV were secured on testing two cast nickel steels made commercially. Attention is directed particularly to reduction of area values as this property is the best indication of the so-called "over-reduced" condition in cast steels.

If the condition of the melt permits use of aluminum without lowering ductility appreciably, it then becomes a useful addition for improving resistance to low temperature impact. Cast steels containing very low sulphur or less than 0.15 per cent carbon usually are not affected adversely by aluminum additions.

## Wrought Steels

Seven heats were melted in the induction furnace and cast into bars approximately 2 inches square and forged down to $\frac{7}{8}$x1$\frac{1}{4}$ inches. Each heat was split so that half was poured after manganese and silicon deoxidation and the remainder treated with 0.08 per cent aluminum in the ladle after the regular deoxidation. All specimens were normalized and drawn at 1200 degrees Fahr. (650 degrees Cent.).

Three steels were made with 0.09 per cent carbon and contained 0, 2 and 3.50 per cent nickel respectively. The remaining steels were made with 0.16 to 0.17 per cent carbon with alloy content the same as in the 0.09 per cent carbon series, except for the fourth steel, which contained 2.75 per cent nickel and thirty points of molybdenum. Room temperature tensile properties and results of Charpy tests using keyhole notch at temperatures down to —200 degrees Fahr. (—130 degrees Cent.) are given in Table V.

Neither of the carbon steels, aluminum-treated or otherwise,

**Table VI**

**Tensile and Low Temperature Impact Properties of Some Commercial Alloy Steel Bars**

| Size | Type Steel | Composition, Per Cent C | Mn | Si | Ni | Mo | Heat Treatment Degrees F. | Mechanical Properties: Yield Point p.s.i., 0.01" extension | Tensile Str. p.s.i. | % Elon. in 2" | % Red. Area | Bhn | McQ-Ehn | Impact, Ft. Lbs., Charpy, Keyhole Notch: Room Temp. | −100° F. | −150° F. | −200° F. |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 1½-inch rounds | Low C, 2% Ni | 0.05 | 0.17 | 0.01 | 2.11 | .... | 1625 AC, 1200 AC | 39,000 | 53,700 | 41.0 | 72.8 | 106 | 1–2 | 57, 59 | 34, 34 | 20, 20 | 3, 5 |
| 3-inch rounds | Low C, 2¼% Ni | 0.13 | 0.44 | 0.20 | 2.69 | .... | 1575 AC, 1200 AC | 53,000 | 71,200 | 36.5 | 72.8 | 144 | 8 | 59, 60 | 37, 39 | 35, 37 | 23, 26 |
| 1¾-inch rounds | Low C, 3¼% Ni, Mo | 0.05 | 0.13 | 0.15 | 3.72 | 0.21 | 1600 AC, 1200 AC | 43,200 | 63,500 | 40.5 | 77.6 | 130 | 1–2 | 70, 71 | 41, 45 | 6, 14 | 4, 5 |
| 2½-inch rounds | Low C, 5% Ni, Mo | 0.09 | 0.86 | 0.22 | 5.35 | 0.34 | 1575 AC, 1200 AC | 51,000 | 108,000 | 26.0 | 65.9 | 217 | 7 | 58, 64 | 35, 41 | 27, 35 | 20, 22 |
| ¼ x 2 inches | Deposited weld metal, Vertical Weld | (0.12) | (0.50) | (0.02) | (2.25) | 0.21 | 1150 AC | 55,700 | 70,700 | 29.7 | 53.9 | 146 | 2–8 | 26, 30 | 11, 22, 26, 26 | 19, 20 | 13, 16 |
| ¼ x 2 inches | Deposited weld metal, Flat Weld | (0.12) | (0.50) | (0.02) | (2.25) | 0.21 | 1150 AC | 52,000 | 68,600 | 34.0 | 64.5 | 139 | 2–8 | 33, 33 | 18, 23 | 14, 21 | 3, 3 |

Numbers in parentheses are estimated values.

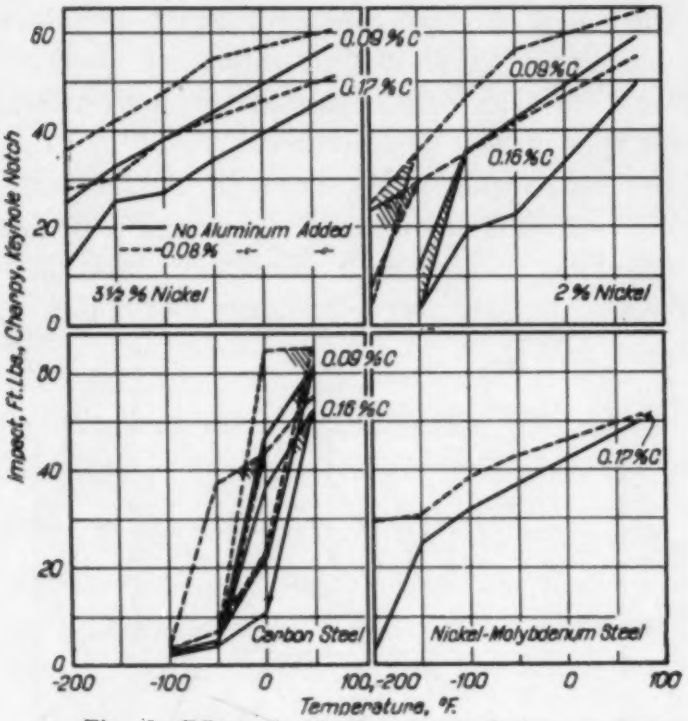


Fig. 5—Effect of Aluminum and Carbon on Low Temperature Impact Properties of Normalized and Drawn Carbon and Nickel Steels.

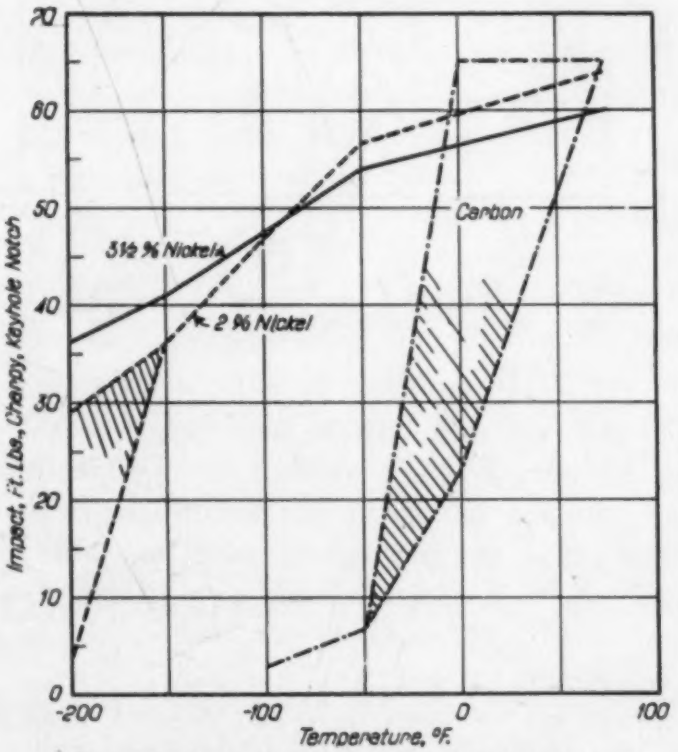


Fig. 6—Effect of Nickel on Low Temperature Impact Properties of Normalized and Drawn 0.09 Per Cent Carbon Steel, Treated With 0.08 Per Cent Aluminum.

resisted embrittlement at —50 degrees Fahr. (—45 degrees Cent.), and there is some question as to their suitability for 0 degrees Fahr. (—18 degrees Cent.) as curves plotted for both steels, Figs. 5, 6 and 7, show a split at this temperature.

The 2 per cent nickel steels, both the 0.09 and 0.16 per cent carbon grades, retained their toughness to —100 degrees Fahr. (—75 degrees Cent.), and with aluminum deoxidation retained good impact resistance down to —150 degrees Fahr. (—100 degrees Cent.) as shown in Fig. 5.

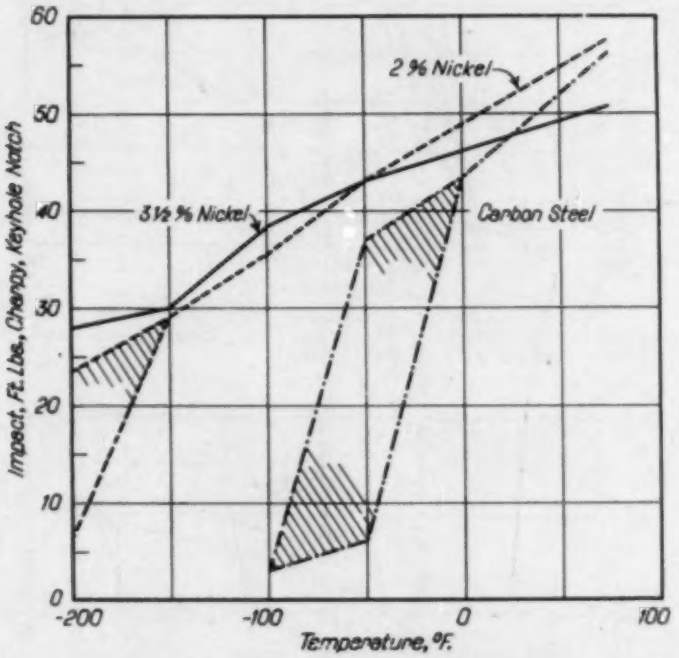


Fig. 7—Effect of Nickel on Low Temperature Impact Properties of Normalized and Drawn 0.16 Per Cent Carbon Steel, Treated With 0.08 Per Cent Aluminum.

The 3.50 per cent nickel steels were tough at all temperatures down to and including —200 degrees Fahr. (—130 degrees Cent.). The 0.09 per cent carbon grade was slightly superior to the 0.17 per cent carbon grade, and the highest values at —200 degrees Fahr. (—130 degrees Cent.) secured with any of these experimental steels were obtained with the 0.09 per cent carbon, 3.50 per cent nickel steel, aluminum-treated.

Good impact values down to —150 degrees Fahr. (—100 degrees Cent.) were secured with the nickel-molybdenum steel with both types of deoxidation, but only the aluminum-treated portion of

this heat gave good values at —200 degrees Fahr. (—130 degrees Cent.).

Bars from four low carbon commercial heats, two of which contained nickel and the remaining two both nickel and molybdenum, were normalized and drawn at 1200 degrees Fahr. (650 degrees Cent.) and tested in impact at temperatures down to —200 degrees Fahr. (—130 degrees Cent.). These bars were treated in the sizes shown in Table VI and specimens were machined out after heat treatment.

It is interesting to note that, in spite of coarse McQuaid-Ehn grain, impact values of the 2 per cent nickel steel at —150 degrees Fahr. (—100 degrees Cent.) are satisfactory and are much superior to those of the experimental 2 per cent nickel steels with similar grain size. The very low carbon content may be responsible.

The 2.75 per cent nickel steel was tough at all temperatures down to —200 degrees Fahr. (—130 degrees Cent.). The point of particular interest in this steel is that treatment was carried out on a 3-inch diameter section.

The 0.05 per cent carbon, nickel-molybdenum steel did not show good toughness at temperatures below —100 degrees Fahr. (—75 degrees Cent.). The McQuaid-Ehn grain size is quite coarse but considering the low carbon content it is somewhat surprising that this steel does not resist embrittlement to lower temperatures. The commercial 2 per cent nickel steel had the same grain size and lower impact resistance at room temperature, yet it would be considered satisfactory for —150 degrees Fahr. (—100 degrees Cent.) service. Without further test data no reason can be advanced for the relatively poor low temperature properties of this steel.

The 0.09 per cent carbon, nickel-molybdenum steel retained its toughness at all temperatures down to —200 degrees Fahr. (—130 degrees Cent.), and the values secured on impact at room and subzero temperatures are quite similar to those of the 2.75 per cent nickel steel.

## Weld Metal Deposited With Nickel Steel Electrodes

As much of the equipment used at low temperatures is fabricated by welding, it was considered advisable to determine the low temperature impact properties of weld metal deposited by steel elec-

trodes of analysis similar to that of rolled material used for this service. Considerable data on low temperature properties of welds have been reported by Aldridge and Shepherd (9) but apparently the temperatures were not carried below —75 degrees Fahr. (—60 degrees Cent.).

Weld deposits ¾x2x8 inches were laid down, in both flat and vertical positions, with a commercial grade of heavily coated $\frac{5}{32}$-inch nickel steel electrodes. These specimens were drawn at 1150 degrees Fahr. (620 degrees Cent.) to simulate a stress relief anneal and then tested at temperatures down to —200 degrees Fahr. (—130 degrees Cent.).

There was some scatter in impact results at temperatures of —100 degrees Fahr. (—75 degrees Cent.) and below, Table VI, but considering the wide range in McQuaid-Ehn grain size (2-8), this would be anticipated. On the whole, impact results of these deposited metal specimens are good down to —150 degrees Fahr. (—75 degrees Cent.).

## Conclusions

Based upon results from testing experimental heats, augmented by tests on specimens representing steel produced commercially, it is concluded that good impact values (Charpy, keyhole notch) can be secured at temperatures down to —200 degrees Fahr. (—130 degrees Cent.) with:

1. Properly deoxidized, normalized and draw cast steel containing a maximum of 0.15 per cent carbon and from 3.50 to 4 per cent nickel, and with cast steels containing a maximum of 0.15 per cent carbon, 2.75 per cent nickel and approximately 0.30 per cent molybdenum.

2. Normalized and drawn wrought steels containing a maximum of 0.17 per cent carbon and a minimum of 2.75 per cent nickel. Optimum nickel content in the plain nickel steels appears to be between 3.50 and 4 per cent.

3. Normalized and drawn wrought steels, containing low carbon and substantial quantities of nickel and molybdenum, that have been deoxidized with aluminum.

4. Any of the above wrought steels after liquid quenching and drawing and, in addition, higher carbon fine-grained nickel,

nickel-molybdenum, and other nickel alloy steels that have been quenched and drawn.

Providing the soundness of welds can be assured by some non-destructive method such as radiographic inspection, welds made with heavily coated, low carbon, 2.50 per cent nickel, 0.20 per cent molybdenum steel electrodes will meet most requirements for impact resistance at temperatures down to —150 degrees Fahr. (—75 degrees Cent.). It is probable that welds made with other steels will maintain good impact resistance at temperatures down to —100 degrees Fahr. (—75 degrees Cent.) and slightly below, but it is believed that there are few steels that will give weld deposits possessing the combination of tensile properties and low temperature impact resistance secured with this nickel-molybdenum steel.

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## DISCUSSION

**Written Discussion:** By S. J. Rosenberg, U. S. Department of Commerce, National Bureau of Standards, Washington, D. C.

It is, perhaps, fortunate that the results secured in impact tests of steels are so little understood that designing engineers still use such results in a

qualitative sense only. Investigators in this field have frequently noted what appear to be erratic and sometimes entirely unexpected results. In impact tests check values usually vary between wide limits and in order to secure what may be considered as "average" values, investigators find it necessary to make several check tests in each case under study.

The practice of evaluating the impact resistance of any material, particularly at low temperatures, from the results of even a comprehensive series of tests upon an individual heat of that material may lead to erroneous conclusions. Presumably check heats of the same steel, made by the same furnace practice, conforming quite closely in chemical composition, having the same heat treatment with approximately the same resultant hardness, can vary quite widely in impact resistance, particularly in that temperature range wherein the impact resistance drops rather rapidly with decreasing temperature.

Some results, as yet unpublished, obtained at the National Bureau of Standards on the effect of grain size and heat treatment upon the impact resistance of S.A.E. 1050 steel at low temperatures indicate that these steels, at least, do not have a characteristic resistance to impact in the same sense, for instance, as they have a characteristic tensile strength. Impact tests made after various heat treatments showed that certain heats consistently had markedly superior impact resistance, particularly at low temperatures, than other heats of the same type steel.

Fine grain size in these steels, as determined by the McQuaid-Ehn test, is no assurance of superiority in impact resistance after heat treatment. On casual reflection this is not surprising since, regardless of the fact that the McQuaid-Ehn grain size may be quite different in any two of these steels, at heat treating temperatures the grain sizes may be quite similar. Of the steels tested, some had a fine and others had a coarse McQuaid-Ehn grain size (the steels were purposely so made). The one particular steel which consistently showed the highest resistance to impact after various heat treatments was a fine-grained steel, but the steel which consistently showed the lowest resistance to impact was also a fine-grained steel. The differences in the impact resistance of these two duplicate heats amounted to about 100 per cent at various test temperatures from +100 to −78 degrees Cent.

The fact that impact tests secured on a single heat of steel are more or less peculiar to that individual heat and will not necessarily be duplicated by another heat of the same type steel has also been noted in steels other than S.A.E. 1050. Individual heats of steel apparently have what may be termed an "inherent" resistance to impact between certain limits and this is dependent upon factors not at present recognized.

**Written Discussion:** By N. A. Ziegler and H. W. Northrup, research laboratories, Crane Co., Chicago.

Messrs. Armstrong and Gagnebin present a very excellent summary of the low temperature behavior of nickel steels. It is quite true that high impact resistance at subzero temperatures can be developed in nickel steels, even containing fairly high percentages of carbon, by quenching and drawing treatment. Our interest, however, regarding this subject is centered primarily on castings, and it is quite well known that any foundryman, whenever possible,

would prefer to avoid quenching treatments. For this reason, our problem is to develop high subzero impact resistance by normalizing and drawing. When this restriction is introduced, carbon content very definitely becomes a limiting factor, which may be judged from the following tabulation of physical properties of a group of experimental steels, each of which was prepared in a 200-pound high frequency furnace and subjected to a triple heat treatment: (1) normalizing from 1750 degrees Fahr., (2) "air quenching" from 1550 degrees Fahr., and (3) drawing at 1200 degrees Fahr.

| Steel No. | Chemical Analysis Si | Mn | Ni | C |
|---|---|---|---|---|
| 1 | 0.26 | 0.70 | 3.61 | 0.03 |
| 2 | 0.23 | 0.72 | 3.61 | 0.07 |
| 3 | 0.21 | 0.50 | 3.62 | 0.15 |

Physical Properties*

| Steel No. | Tensile Test Bars, Room Temperature: Tensile Strength p.s.i. | Yield Point p.s.i. | Prop. Limit p.s.i. | Elong. Per Cent | Red. of Area Per Cent | Impact Resistance (Ft. Lbs.): Room Temp. | 0 | −50 | −100 | −150 | −175 Degrees Fahr. |
|---|---|---|---|---|---|---|---|---|---|---|---|
| 1 | 62,800 | 44,000 | 42,000 | 43.0 | 78.7 | 58.5 | 58.5 | 60.8 | 59.2 | 53.2 | 41.5 |
| 2 | 68,300 | 50,000 | 50,000 | 40.5 | 75.9 | 52.3 | 49.2 | 54.3 | 39.3 | 35.8 | 33.0 |
| 3 | 75,800 | 58,000 | 57,500 | 35.0 | 57.0 | 25.7 | 27.0 | 19.3 | 18.5 | 10.8 | 10.3 |

*Each physical property is an average of several tests.

In order to obtain good impact resistance in a 3.5 to 4.0 per cent nickel steel at temperatures under −100 degrees Fahr., carbon content should be restricted to 0.07, and preferably to 0.05 per cent. The tensile strength of such steels is below 70,000 pounds per square inch, a figure which is generally desired, but it seems that the latter has to be sacrificed for developing high subzero impact resistance.

Attempts have been made to strengthen nickel steels by alloying them with small amounts of vanadium. Physical properties of such steels, prepared in a similar manner, and subjected to a similar triple heat treatment, are given in the following tabulation.

| Steel No. | Chemical Analysis Si | Mn | Ni | V | C |
|---|---|---|---|---|---|
| 1 | 0.27 | 0.65 | 3.55 | 0.13 | 0.02 |
| 2 | 0.27 | 0.64 | 3.63 | 0.15 | 0.07 |
| 3 | 0.18 | 0.65 | 3.71 | 0.12 | 0.14 |

Physical Properties*

| Steel No. | Tensile Test Bars, Room Temperature: Tensile Strength p.s.i. | Yield Point p.s.i. | Prop. Limit p.s.i. | Elong. Per Cent | Red. of Area Per Cent | Impact Resistance (Ft. Lbs.): Room Temp. | 0 | −50 | −100 | −150 | −175 Degrees Fahr. |
|---|---|---|---|---|---|---|---|---|---|---|---|
| 1 | 69,650 | 45,500 | 38,000 | 36.0 | 76.0 | 46.3 | 48.7 | 37.0 | 37.0 | 31.5 | 29.0 |
| 2 | 84,550 | 65,000 | 62,000 | 29.5 | 68.0 | 44.3 | 33.7 | 18.9 | 22.7 | 17.5 | 15.8 |
| 3 | 99,250 | 84,500 | 82,000 | 24.0 | 55.0 | 23.8 | 16.7 | 12.0 | 12.7 | 7.7 | 6.3 |

*Each physical property is an average of several tests.

It is true that vanadium does strengthen steels of this type, but at the same time it makes them, contrary to our expectations, more brittle. The straight nickel steel with 0.07 per cent carbon (first tabulation) has a very

satisfactory impact resistance at temperatures below —100 degrees Fahr., while a similar steel, but containing 0.15 per cent vanadium (second tabulation), at the same temperatures is rather brittle.

Indications are that molybdenum additions, in this sense, may be more satisfactory. Only one such composition, similarly prepared and heat treated, has been tried so far.

| Chemical Analysis | | | | |
|---|---|---|---|---|
| Si | Mn | Ni | Mo | C |
| 0.40 | 0.71 | 2.86 | 0.35 | 0.12 |

| Physical Properties* | | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| Tensile Test Bars, Room Temperature | | | | | Impact Resistance (Ft. Lbs.) | | | | | |
| | | | | Red. of | | Degrees Fahr. | | | | |
| Tensile Strength p.s.i. | Yield Point p.s.i. | Prop. Limit p.s.i. | Elong. Per Cent | Area Per Cent | Room Temp. | 0 | —50 | —100 | —150 | —175 |
| 83,800 | 62,500 | 51,500 | 31.0 | 67.0 | 39.2 | 37.5 | 32.0 | 26.4 | 24.0 | 20.2 |

*Each physical property is an average of several tests.

Physical properties of this steel are very good, and the subzero impact resistance is satisfactory. Nevertheless, the latter is considerably below the impact resistance of the lowest carbon straight nickel steel (first tabulation). It may be that by reducing carbon, the low temperature impact resistance of the nickel-molybdenum steel could be improved. At the same time molybdenum may act as a strengthener, thus improving its tensile strength. This, however, remains to be seen. In a word, very low carbon is a rather necessary condition for good impact resistance, at temperatures below —100 degrees Fahr. in all types of low alloyed cast nickel steels.

## Oral Discussion

C. E. SIMS:[2] There is only one point I wish to discuss. If I understood Mr. Armstrong correctly I want to take exception to his statement in connection with the use of aluminum in cast steel. He said that aluminum is liable to cause low ductility in cast steel and that unless the steel is properly prepared first, as for example by severe oxidation, then aluminum had best be omitted.

The manner in which aluminum can be used without detriment to the ductility of cast steel was described in a paper[3] read before the American Foundrymen's Association last year. It is apparently true that steel for castings in general is benefited by an oxidation treatment which entails a vigorous boil, but our experience indicates that, as far as the use of aluminum is concerned, the preparation of the steel is of minor importance as compared to using the correct amount of aluminum. There is no inference, of course, that aluminum will improve the ductility of a steel otherwise low in ductility.

E. E. THUM:[4] A question was asked me recently by a rather prominent construction engineer who was designing some large equipment for low temperature work. In the A.S.M. Metals Handbook he found that certain alloy

[2]Research metallurgist, Battelle Memorial Institute, Columbus.

[3]"Effect of Aluminum on the Properties of Medium Carbon Steel," C. E. Sims and F. B. Dahle, *Transactions*, American Foundrymen's Association, 1938.

[4]Editor, METAL PROGRESS, Cleveland.

steels were recommended for such duty because of their high impact strength at low temperature. His question was:

"What does this high impact strength mean, and is there a body of experience which indicates what impact strength the metal in a pressure vessel operating at minus 100 degrees Fahr. should have?" I said that I did not know of any published data on this point, or any information that difficulties had been experienced in practice with metals having, say, less than 15 foot-pounds impact at working temperature, but seemingly those who should know believe that high impact strength is a desirable characteristic—possibly an additional margin of safety. His question "How much extra would I be warranted in paying for high impact strength?" I had to leave unanswered.

So I wish merely to ask those who are making such low temperature equipment (valves, pipe and pressure vessels) to clarify that point if they can. Is there any bad experience recorded with steels which have low impact strength at low temperature? If not, why should there be such an insistence for high impact strength at low temperature? Railroads get along pretty well with rail steel which has not a very high impact strength even at temperate temperatures.

### Authors' Reply

The data submitted in the discussion by Messrs. Ziegler and Northrup added considerably to the value of the paper. On the whole their results confirm those of the authors', the only disagreement being the allowable tolerance for carbon. Attention is directed to heats F and L in Tables I and II which had 0.15 and 0.13 per cent carbon respectively. Both steels had satisfactory impact properties at —150 degrees Fahr., even though heat F, a commercial melt, is considered somewhat under-alloyed for such low temperature.

Mr. Rosenberg calls attention to lack of consistency in impact properties, both at room temperatures and at subzero temperatures, among several steels of the same approximate analysis having approximately the same McQuaid-Ehn grain size. It is significant that the steels investigated by Mr. Rosenberg contained appreciably more carbon (S.A.E. 1050) than the steels reported in the paper and it is quite probable that these higher carbon steels are more susceptible to manufacturing variables than steels composed predominantly of ferrite. The authors do agree, however, that the present state of the art does not permit use of composition, grain size and heat treatment as criteria for resistance to impact, which indicates the advisability of subjecting each lot to test, particularly when service involves temperatures below —100 degrees Fahr. But the same qualifications hold for the static properties as knowledge of composition, grain size and heat treatment is also not considered sufficient to preclude use of the conventional tensile test in determining strength properties.

The authors agree that use of correct amount of aluminum is important, but it appears that Mr. Sims' statements are somewhat contradictory in the following: "It is apparently true that steel for castings in general is benefited by an oxidation treatment which entails a vigorous boil, but our experience indicates that as far as the use of aluminum is concerned the preparation of steel is of minor importance as compared to using correct amount of aluminum.

There is no inference, of course, that aluminum will improve the ductility of a steel otherwise low in ductility."

The authors fail to understand why preparation of steel is of minor importance in conjunction with use of aluminum if aluminum addition fails to improve ductility when this property is obviously low. Until more conclusive evidence can be presented, it is believed that the practice of melting steel under oxidizing conditions will be the accepted method for insuring good ductility in cast steel, regardless of whether aluminum is used or not.

In answer to questions raised by Mr. Thum, there have been failures in oil refineries directly traceable to materials having low impact resistance at subzero temperatures. As to how much extra cost is justified for high impact properties, it depends not only upon replacement cost but extent of damage such failures might incur. If a fire results, which has happened in refineries because of such failures, the extra cost for materials of high impact strength will be only a few hundredths of a per cent of the cost of the damage.

It is quite true railroads get along pretty well with steel rails which have comparatively low impact strength even at temperate temperatures. It might also have been pointed out that thousands of chill cast iron wheels have been used on freight cars. However, the Northern railroads do recognize the embrittling effect of low temperature and use materials in highly stressed working parts of locomotives and passenger cars that possess good low temperature properties.

One manufacturer of crushers, because of past experience, equips all crushers that are to operate in extremely cold climates such as encountered in the Northwest United States and in Canada with shafts of a steel that possess good impact properties at low temperatures. This is cited simply as another example of a practical application.

# STRESS CORROSION CRACKING OF THE AUSTENITIC CHROMIUM-NICKEL STEELS AND ITS INDUSTRIAL IMPLICATIONS

By James C. Hodge and John L. Miller

*Abstract*

*A number of service failures of the austenitic chromium-nickel steels occurred as a result of the formation of localized cracks. Service conditions had not involved repeated or variable stressing and had not necessarily involved materials susceptible to intergranular corrosion. Investigation disclosed that the failures had involved a condition of internal static stress and exposure of the stressed material to certain corrosive media, which in some cases were so inactive that they left ordinary carbon steel unaffected.*

*The failures were identified as being of the stress corrosion type and closely allied to the "season cracking" of certain brasses and to the stress corrosion cracking exhibited less frequently in other metals, as for example the caustic embrittlement of boiler plate. It was determined that such stress corrosion cracking in the austenitic chromium-nickel steels may manifest itself either as an intercrystalline or as a transcrystalline failure. Factors which include the stress intensity, strength of the corrosive medium and the susceptibility of the material to intercrystalline corrosion determine the path of the failure.*

*Stabilizing elements, such as titanium and columbium, were found to be ineffective in preventing transcrystalline stress corrosion cracking, but intercrystalline susceptibility was absent and a somewhat higher stress value was required. To eliminate the possibility of stress corrosion failure, fabricating stresses must be removed or reduced by an elevated temperature treatment and the material slowly cooled to prevent thermal stressing. The evidence indicates that even such heat treatment is ineffective when two metals having different coefficients of thermal expansion, such as 18-8 and carbon steel, are joined by welding.*

*The corrosive media capable of producing stress corrosion cracking in the austenitic chromium-nickel steels are limited in number. The mixture of ethyl chloride and*

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*water was found to be very effective in producing stress corrosion failure. Ferric chloride and mercuric chloride solutions were also found to have a very damaging effect.*

*A discussion is presented in which it is argued that the mechanism of stress corrosion failure is similar to that of corrosion fatigue cracking.*

## Introduction

THE austenitic chromium-nickel steels, containing an approximate minima of 18 per cent chromium and 8 per cent nickel, are increasingly favored for industrial processing services. Although initially costly, such alloys are economically utilized because of the embodiment of excellent mechanical properties with resistance to a wide variety of corrosive influences at low and high temperatures. One serious defect, that of carbide precipitation with accompanying grain boundary chromium impoverishment and with consequent susceptibility to intercrystalline corrosion, was early recognized and methods developed for elimination of the occurrence or nullification of its effects. This was accomplished by employing high temperature quenching treatments to retain the alloy in a single phase but unstable condition, or by the use of stabilizing elements, particularly titanium or columbium. These elements when present in sufficient amount preferentially unite with the carbon of the alloy permitting the chromium to remain in solid solution even when the thermal environment is temporarily or permanently in the carbide precipitation range. If such measures are not adopted the alloy is liable to failure by intercrystalline disintegration in services involving contact with strong corrosive media and the makers, fabricators and users of such steels are aware of such a possibility and on guard to prevent its occurrence.

In common with other metals and alloys these austenitic stainless steels possess relatively low corrosion-fatigue values. By reason of their relatively high coefficient of thermal expansion, high stress levels may be more rapidly induced in high temperature process equipment subject to variable service temperatures and fatigue and corrosion-fatigue effects are possibly of greater importance in the case of these steels than in ferritic steels. Many failures of these austenitic steels by fatigue or corrosion fatigue, particularly in heat exchangers, have been experienced. Typical failures by corrosion

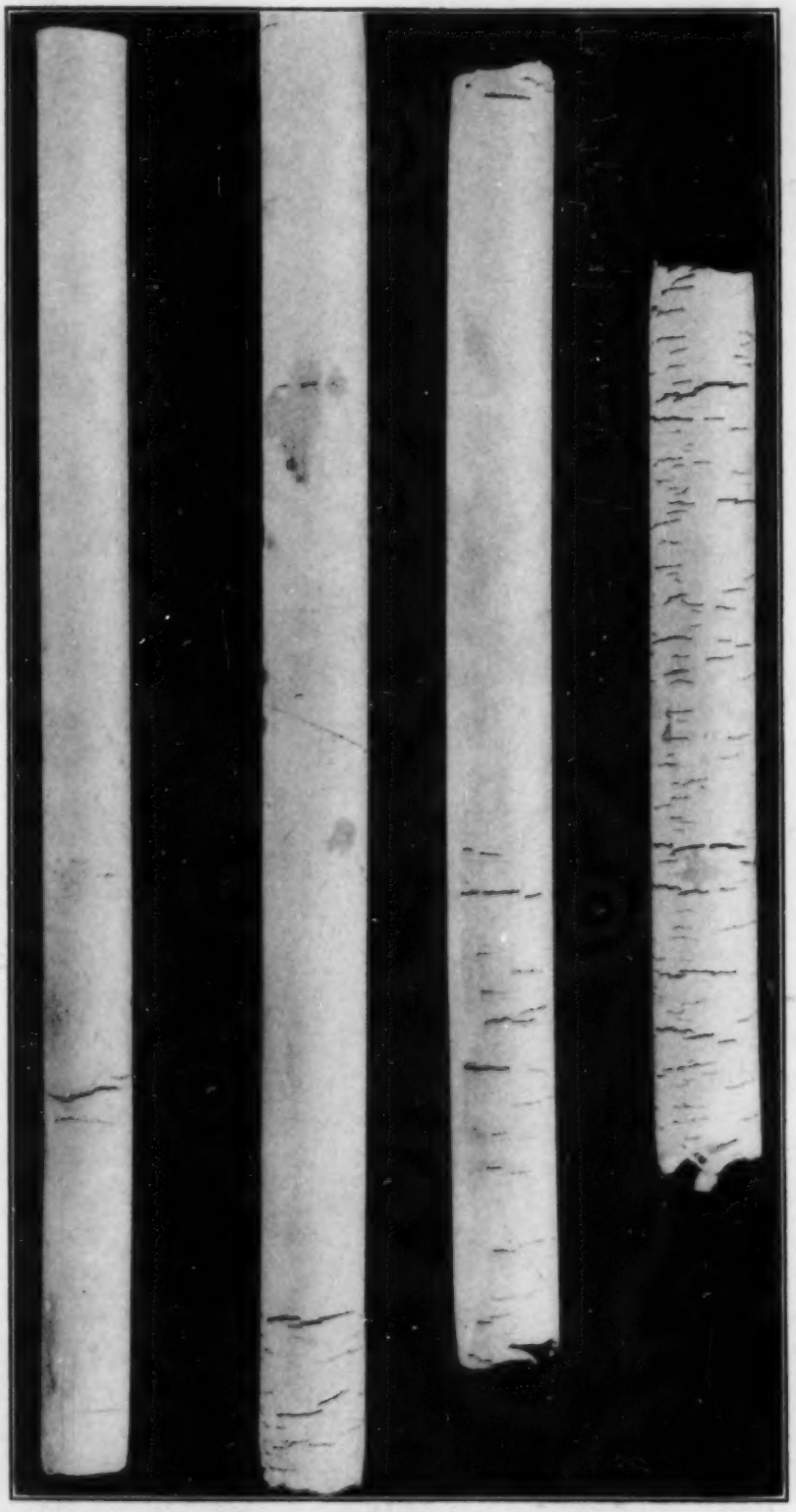

Fig. 1—Fatigue Cracks in 18-8 Tubes, Which Operated in Condenser Units Which Did Not Provide Adequate Freedom for Expansion of the Tubes. (Repickled to Show Cracks). Naphtha Outside of Tube, Brackish Water Inside. (H. D. Newell, The Babcock and Wilcox Tube Company).

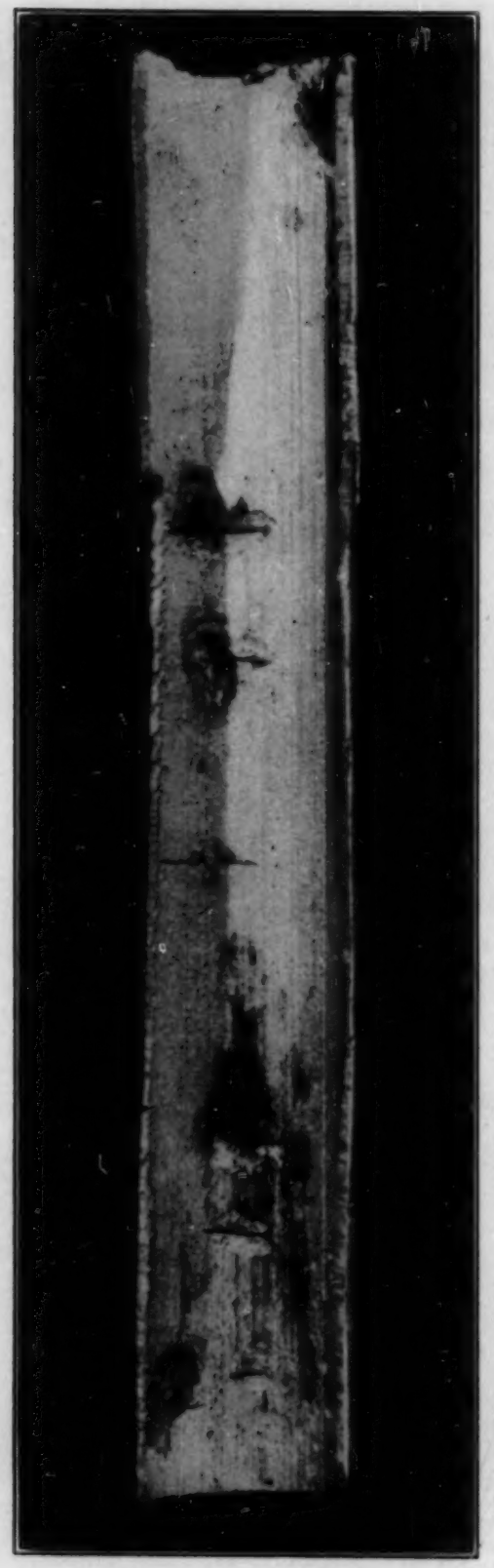

Fig. 2—Internal Pitting of 18-8 Condenser Tube with Corrosion Fatigue Cracking Occurring at Bottom of Pits. Tube Operated in Contact with Brackish Water. (H. D. Newell, The Babcock and Wilcox Tube Company).

fatigue of condenser and exchanger tubes of austenitic 18-8 alloy steel are illustrated in Figs. 1 and 2. Cracks are normal to the direction of the service stresses and in tube material heat treated by quenching from high temperatures the paths of the cracks are transcrystalline. Designers of such equipment now recognize this danger and generally provide for the free expansion of the austenitic steel parts.

Within recent years, however, an increasing number of failures of these austenitic chromium-nickel steels have occurred under circumstances not involving repeated or variable stressing and not necessarily involving material susceptible to intergranular corrosion. Such failures do involve a condition of internal static stress and exposure of the stressed material to certain corrosive media which, in some cases at least, are so inert that they leave ordinary carbon steel relatively unaffected over long exposure periods. In contrast with the general overall corrosion attack characterizing failures due to intergranular chromium impoverishment but in common with failures by fatigue or corrosion-fatigue effects, such failures are of localized character, i. e., cracks.

The localized nature of the failures suggests that stress is the primary requisite of the occurrence. The term "stress corrosion" has properly been applied to the phenomenon. This expression has been used in preference to the older but less descriptive term "season cracking" as applied to like failures in the brasses.

For several years since the first serious failure of this type was brought to our attention the authors of this paper have been continually engaged in a study of the circumstances surrounding such phenomena and a considerable amount of experimental work has been devoted to an endeavor to determine quantitatively the stress values and the corrosive media necessary for the production of such failures.

These localized failures have been found to characterize austenitic chromium-nickel steels under certain critical circumstances, although the manner of cracking may vary, dependent upon several factors, including stress intensity and strength of the corrosive medium. It has been determined that such stress corrosion failures may be of an intercrystalline nature, of a transcrystalline nature, or in a given installation, and even in a single specimen, a combination of the two. Generally, a localized stress corrosion failure will be intercrystalline if the metal is susceptible to intercrystalline corro-

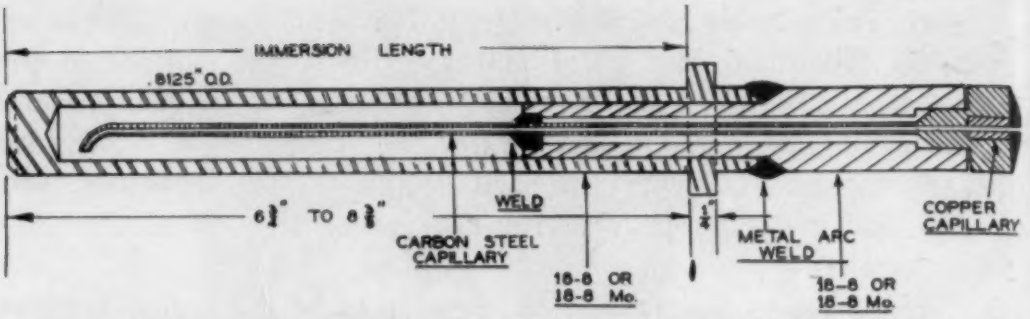


Fig. 3—Sketch Showing the Usual Construction of the Stainless Steel Thermowells.

sion. For high stress values or when the intercrystalline susceptibility is slight, failure may readily be transcrystalline.

Alloys immune to intercrystalline corrosion, such immunity being obtained by a high temperature quenching treatment, or because of the presence of a stabilizing element such as titanium or columbium, fail in a transcrystalline manner. In either case, whether fracture is intercrystalline or transcrystalline, no measurable elongation or reduction of area will take place, although the alloys are extremely ductile.

## Service Failures

In order to clarify exposition of the scope of the investigation, two case histories will be cited. These cases are typical of several service failures which have been encountered and which may be grouped into two general classifications: one in which the alloys were resistant to intercrystalline corrosion and the failures transcrystalline, and the other in which the alloys were not resistant to intercrystalline corrosion and the failures were intercrystalline.

*Case No. 1: Stainless Steel Thermowell Failures.* In many plants manufacturing paper using the sulphite process, the sulphite liquor is continuously circulated between the pulp digester and an external heat exchanger. This heat exchanger and the piping to and from the digester are made using the austenitic chromium-nickel steel to withstand the corrosive action of the sulphite liquor. To permit measurement and control of the liquor temperature, thermowells using ethyl chloride ($C_2H_5Cl$) under pressure were installed in the pipe lines.

A typical thermowell assembly is shown in Fig. 3. The first thermowells used were machined from forged and water-quenched 18 per cent chromium-8 per cent nickel austenitic steel rounds, the parts welded together and the thermowells installed without sub-

Fig. 4—View Showing an Inside Section of an 18 Per Cent Chromium, 8 Per Cent Nickel, 3 Per Cent Molybdenum Austenitic Steel Thermowell After Failure in Service. This Thermowell Had Been Water-Quenched From 2000 Degrees Fahr. All Cracks are Transcrystalline and Begin at the Inside Where the Metal Surface was in Contact With Ethyl Chloride.

Fig. 5—View of a Typical Superheater Section Ready for Installation.

sequent heat treatment. Failure by cracking occurred after service lives varying from three days to one year. The cracks were transcrystalline and invariably began at the inside where the metal was in contact with the ethyl chloride, a supposedly noncorrosive gas. The carbon steel capillary tubes remained in perfect condition and the outsides of the thermowells, which were in contact with the corrosive sulphite liquor, were in good condition.

The first recommendation made to avoid failure was to water-quench the thermowells after fabrication in order to eliminate the effects of machining and welding. These water-quenched thermowells rapidly failed in service, a typical failure being shown in Fig. 4. At this stage the investigation became a laboratory problem and the procedure followed and results secured will be discussed later.

*Case No. 2: Stainless Steel Superheater Tube Failures.* Superheaters are used to increase the temperature of steam in a boiler before it passes to the exit. They consist essentially of closely spaced rows of tubes placed in the path of the heated gases as they issue from the combustion chamber. In modern high capacity units the exit steam temperature may be of the order of 950 degrees Fahr. necessitating the use of alloy tubes to withstand the creep stresses and the oxidizing influences of the steam and furnace gases.

The austenitic chromium-nickel steels, particularly 18-8, have been used in the highest temperature regions since 1928. The usual

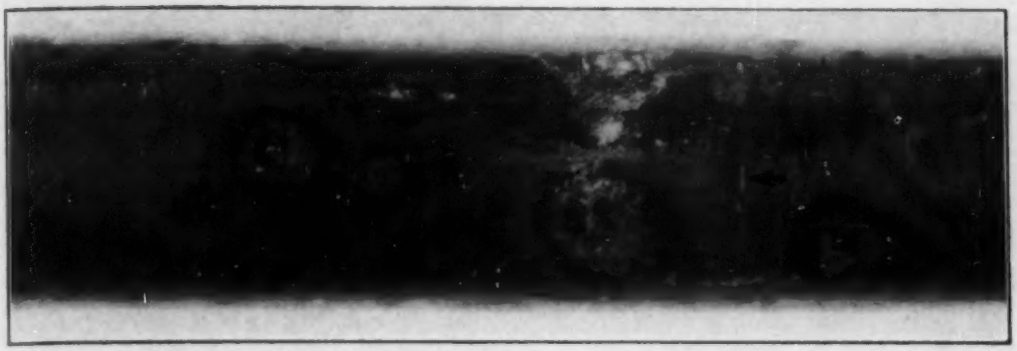

Fig. 6A—Outside View of a 2¼ Inches O.D., 18 Per Cent Chromium, 8 Per Cent Nickel Austenitic Steel Tube After Service Failure. Note the Two Circumferential Cracks. These Cracks Had Their Inception at the Inside Tube Surface and are Intercrystalline.

Fig. 6B—Inside Views of Three 18 Per Cent Chromium, 8 Per Cent Nickel Austenitic Steel Tubes After Service Failures. The Cracking is Intercrystalline and Invariably Began at the Inside Surfaces.

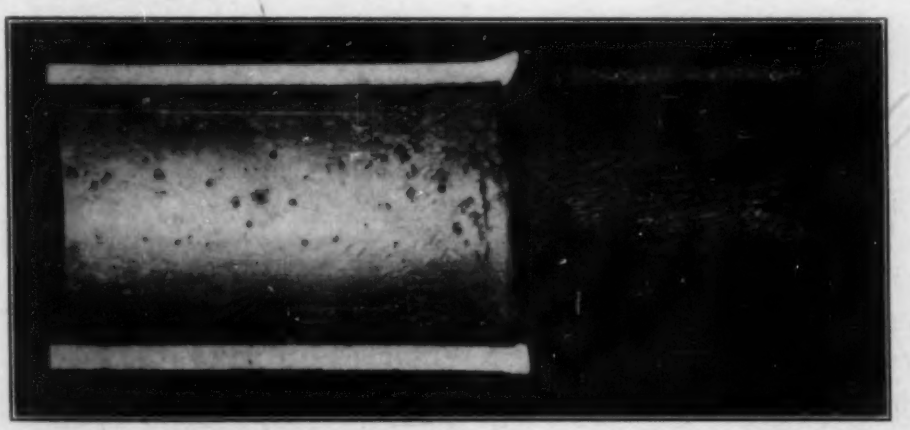

Fig. 6C—Inside View of a Section Through a Flash Weld Joining an 18-8 Tube (Left) and a Carbon Steel Tube (Right). Cracking Has Occurred in the 18-8 Portion.

method of fabrication has been to flash butt weld sections of water-quenched tubes together to form long lengths and then hot bend the

tubes to secure sections such as illustrated in Fig. 5. The actual tube diameters and thicknesses will vary considerably, but in the usual stationary boiler will be 2 to 2½ inches O.D. by $\frac{3}{16}$ inch thick. Ordinary procedure has been to place the 18-8 sections so fabricated in service without heat treatment. These 18-8 tubes operate within a temperature range where carbide precipitation occurs and the metal becomes susceptible to intercrystalline corrosion. General intercrystalline corrosion does not occur over long service periods, but as time of service is prolonged isolated intercrystalline cracking has been detected in several superheater installations so fabricated. The cracking begins at the inside of the tube which is in contact with the steam. Typical illustrations of such failures are shown in Fig. 6.

The first serious failure of this type was encountered about four years ago in a superheater constructed using 18-8, 5 per cent chromium-0.5 per cent molybdenum steel, and carbon steel in various parts of the sections. Because of the presence of a hard and brittle region in the 5 per cent chromium-0.5 per cent molybdenum steel at and adjoining the flash resistance butt weld between the 18-8 and the 5 per cent chromium-0.5 per cent molybdenum steel, corrective heat treatment was given the sections after fabrication.

This heat treatment was at 1350 degrees Fahr. (730 degrees Cent.) for one hour followed by furnace cooling and constituted the one change in practice from that usually employed. It was realized that the temperature of 1350 degrees Fahr. (730 degrees Cent.) was well within the carbide precipitation range of approximately 800 to 1650 degrees Fahr. (425 to 900 degrees Cent.) but in view of the fact that the metal operating service temperature of 1050 to 1100 degrees Fahr. (565 to 595 degrees Cent.) was also within this range, as were the service temperatures of other previously fabricated units, no complications were anticipated. Failure by localized intercrystalline cracking of the 18-8 portion of the superheater developed in two tubes soon after the boiler was placed in service. These sections were replaced but failures continued to develop and within nine months it was necessary to remove the 18-8 portion of the superheater. All cracks were localized and intercrystalline, developing in the 18-8 near the weld joining the 18-8 and the 5 per cent chromium-0.5 per cent molybdenum steel tubing or at locations where the 18-8 tubing had been cold bent. All cracks started on the inside (steam side) of the tubes. The 5 per cent chromium-

0.5 per cent molybdenum steel tubes and the carbon steel tubes had been unaffected by the service. Experimental work conducted in an effort to explain the causes of this and other similar failures is detailed in the following:

*Stress Corrosion Cracking by Ethyl Chloride—Thermowells.* The fact that the cracks in the thermowells had originated on the inside surface in contact with ethyl chloride indicated this compound to be the active agent.

All the thermowells that failed in service were analyzed chemically and found to be either 18-8 or 18-8 molybdenum. Two typical analyses are shown in Table I.

**Table I**

| Element | 18-8 Molybdenum $KA_2Mo$ | 18-8 $KA_2S$ |
|---|---|---|
| Carbon | 0.10 | 0.07 |
| Manganese | 0.46 | 0.45 |
| Silicon | 0.57 | 0.64 |
| Chromium | 17.85 | 17.96 |
| Nickel | 9.60 | 9.22 |
| Molybdenum | 2.87 | .... |

Several samples of the ethyl chloride were analyzed, and small amounts of water detected in each sample. With water present in the ethyl chloride, decomposition in accordance with the following chemical equation is possible.

$$2C_2H_5Cl + H_2O \rightarrow C_2H_5OH + 2HCl + C_2H_4$$

To determine if the localized cracking was due to the presence of small amounts of decomposition products such as hydrogen chloride or nascent chlorine resulting from reaction between the ethyl chloride and water two 18-8 molybdenum thermowells were fabricated and then heat treated by water-quenching from 2000 degrees Fahr. (1095 degrees Cent.). They were then subjected to test under conditions simulating those of service for a period of three weeks. One of the thermowells contained ethyl chloride to which was added about 0.5 per cent of water. The other thermowell contained ethyl chloride from which all water had been abstracted. After test, the thermowells were sectioned and the one which had been exposed to dry ethyl chloride showed no evidence of cracks (Fig. 7A).

The second thermowell which contained ethyl chloride plus water presented an internal appearance as shown in Fig. 7B.

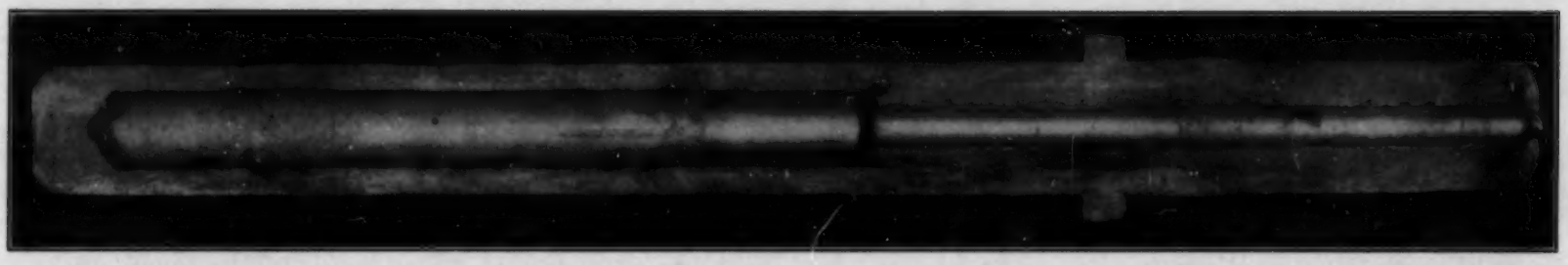

Fig. 7A—Inside View of a Longitudinal Half Section of the 18 Per Cent Chromium, 8 Per Cent Nickel, 3 Per Cent Molybdenum Steel Thermowell After Exposure to the Action of Water-Free Ethyl Chloride for a Period of 21 Days. The Thermowell Had Been Water-Quenched From 2000 Degrees Fahr. Before Test. No Cracking Occurred.

Fig. 7B—Inside View of a Longitudinal Half Section of the 18-8 Molybdenum Steel Thermowell After Exposure to the Action of Ethyl Chloride Plus Water for a Period of 21 Days. The Thermowell Had Been Water-Quenched Before Test. The Cracks are Very Numerous and All are Transcrystalline.

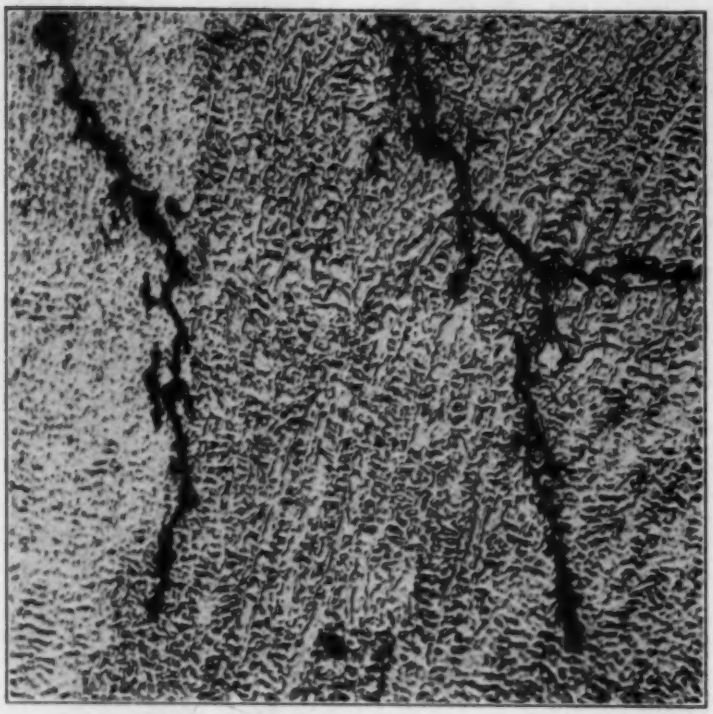

Fig. 8A—Photomicrograph of Cracks Found in the Thermowell of Fig. 7B. Transcrystalline Cracks in Weld Metal. 200 Diameters.

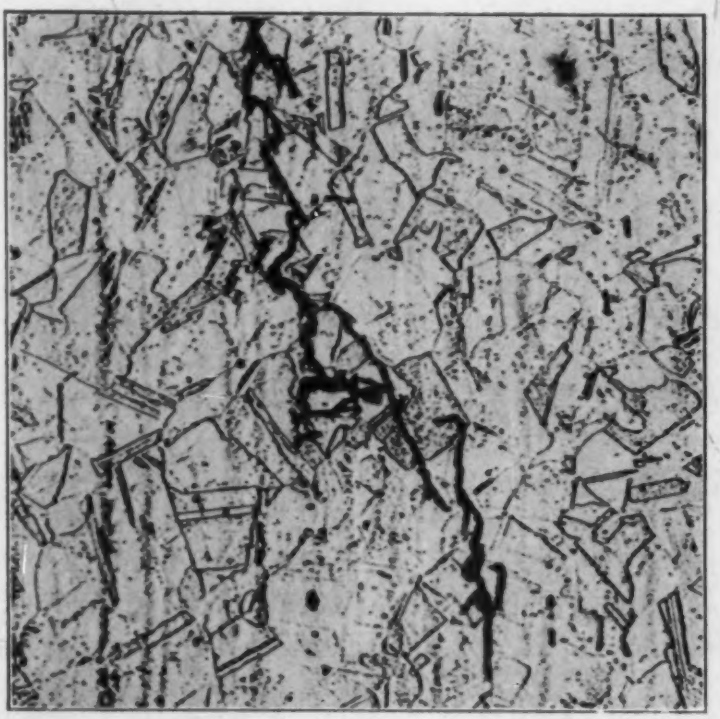

Fig. 8B—Transcrystalline Cracks Through the Tube Material.

Cracks were found that extended completely through the tube wall, and there were thousands of other cracks present that were partially through. Microscopical examination revealed that these cracks, in common with all others found in other thermowells, were transcrystalline in nature. In Fig. 8 there are presented at 200 diameters magnification views of three of the transcrystalline cracks found in this specimen. A section of this failed tube was tested

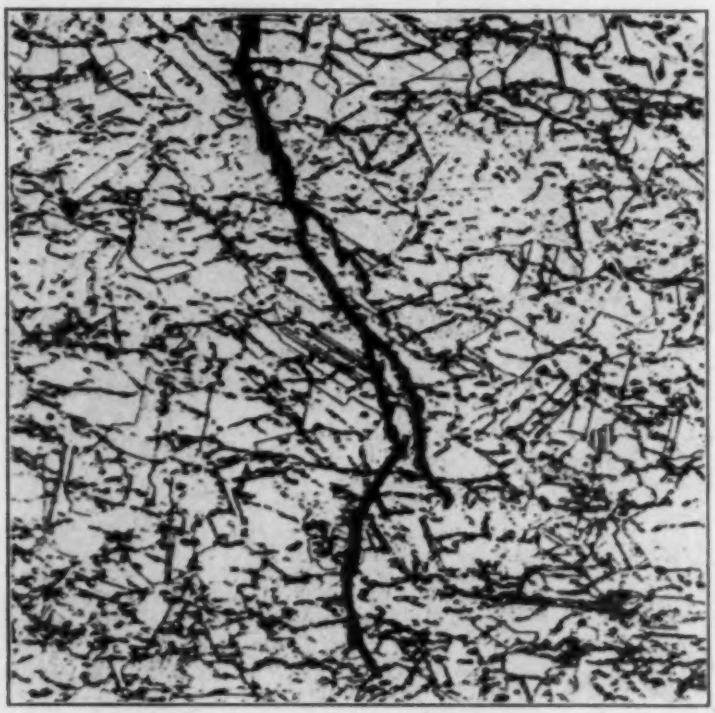

Fig. 8C—Transcrystalline Cracks Through the Tube Material.

by boiling in a water solution containing 10 per cent sulphuric acid and 3 per cent copper sulphate for 100 hours. Bending of this section of the tube with the outside of the tube forming the outside of the bend indicated, as expected, the absence of any tendency to intercrystalline attack.

The comparative behavior of these two identical thermowells in the dry and water-containing ethyl chloride indicated that the active corrosive medium was hydrochloric acid or some other decomposition product, possibly nascent chlorine. The stress factor in the failure of the second experimental thermowell was suspected as being due to the internally stressed condition resulting from differences in cooling rates of inside and outside of the thermowell during the water-quenching treatment. To determine the various

stressed conditions contributing to failure, additional experimental thermowells of 18-8 molybdenum material were constructed as follows:

A. Machined and weld-assembled. This thermowell contained machining and welding stresses and was, undoubtedly, susceptible to intercrystalline corrosion at regions adjoining the weld.

B. Machined, weld-assembled and water-quenched from 2000 degrees Fahr. (1095 degrees Cent.). This thermowell contained stresses resulting from differences in cooling rates between inside and outside surfaces, but was not susceptible to intercrystalline corrosion.

C. Machined, weld-assembled and annealed at 1550 degrees Fahr. (845 degrees Cent.) for 72 hours, followed by slow furnace cooling. The purpose of the prolonged annealing period was to secure precipitation and agglomeration of the carbides and migration of chromium to the areas depleted by the formation of chromium carbide, $Cr_4C$. The purpose of the slow furnace cooling was to eliminate the possibility of forming new stressed regions which might have resulted from differential cooling rates. This thermowell was free of susceptibility to intercrystalline corrosion and was also stress free.

The three thermowells were then tested for two weeks in contact with ethyl chloride containing 0.5 per cent of water to insure that decomposition of the ethyl chloride would occur. The results secured are indicated in Fig. 9.

Thermowell A failed through the weld but cracking also developed in the body of the tube. The cracks were extremely fine and were not discernible in the macro view. The photomicrograph of Fig. 9A shows the slightly elongated condition of the grains in the body of the tube and the transcrystalline nature of the cracking. This thermowell was not heat treated and the grain structure shown is that of the material in the as-received condition. Failure resulted from the presence of welding and machining stresses added to the stress, if any, present in the as-received material.

Thermowell B (Fig. 9B) cracked badly because of the presence of thermal stresses resulting from rapid cooling from high temperature, but thermowell C (Fig. 9C), being in a stress-free condition, was unaffected. All cracks in Thermowells A and B were transcrystalline.

*Stress Corrosion Cracking by Ethyl Chloride—18-8 Superheater Tubing.* The experimental work on thermowells described in the previous section had shown that a mixture of ethyl chloride and water was very effective in producing cracking in stressed austenitic chromium-nickel steels. Other tests using this agent were

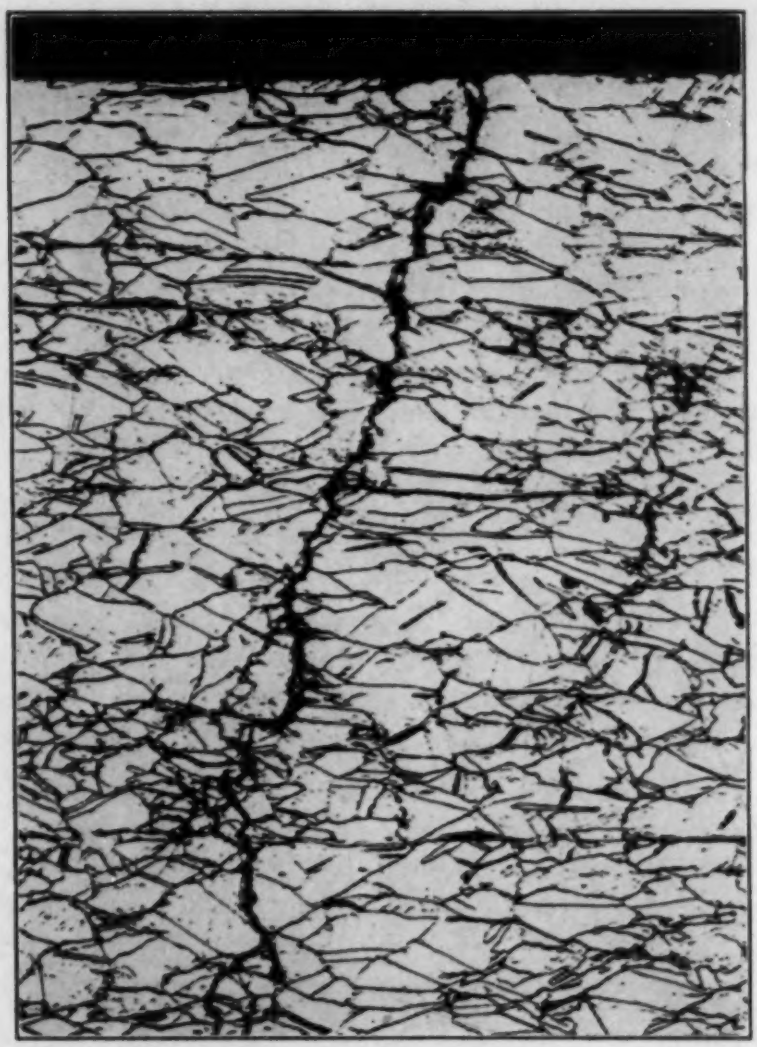

Fig. 9A—Photomicrograph of the Body 18 Per Cent Chromium, 8 Per Cent Nickel, 3 Per Cent Molybdenum Experimental Thermowells After Testing for 14 Days in Ethyl Chloride Plus 0.5 Per Cent of Water of the Tube. All Cracks were Transcrystalline. × 200.

undertaken to determine the susceptibility to stress corrosion cracking of the 18-8 tubing of the failed superheater previously referred to. The tests involved determination of the relative susceptibility of this material as affected by previous thermal history and other fabrication and construction factors. The following gives the fabrication procedure, heat treatment and the results secured in the

Fig. 9B—Machined, Weld Assembled and Water-Quenched From 2000 Degrees Fahr. All Cracks were Transcrystalline.
Fig. 9C—Machined, Weld Assembled and Annealed at 1550 Degrees Fahr. For 72 Hours and Slowly Cooled. No Cracking Occurred.

Fig. 10—View Indicating Construction of the Bottles Used in the Stress Corrosion Experiments Involving Ethyl Chloride. The One Shown is Bottle "C" Which Did Not Crack. The Bottle Was Sectioned Before Photographing.

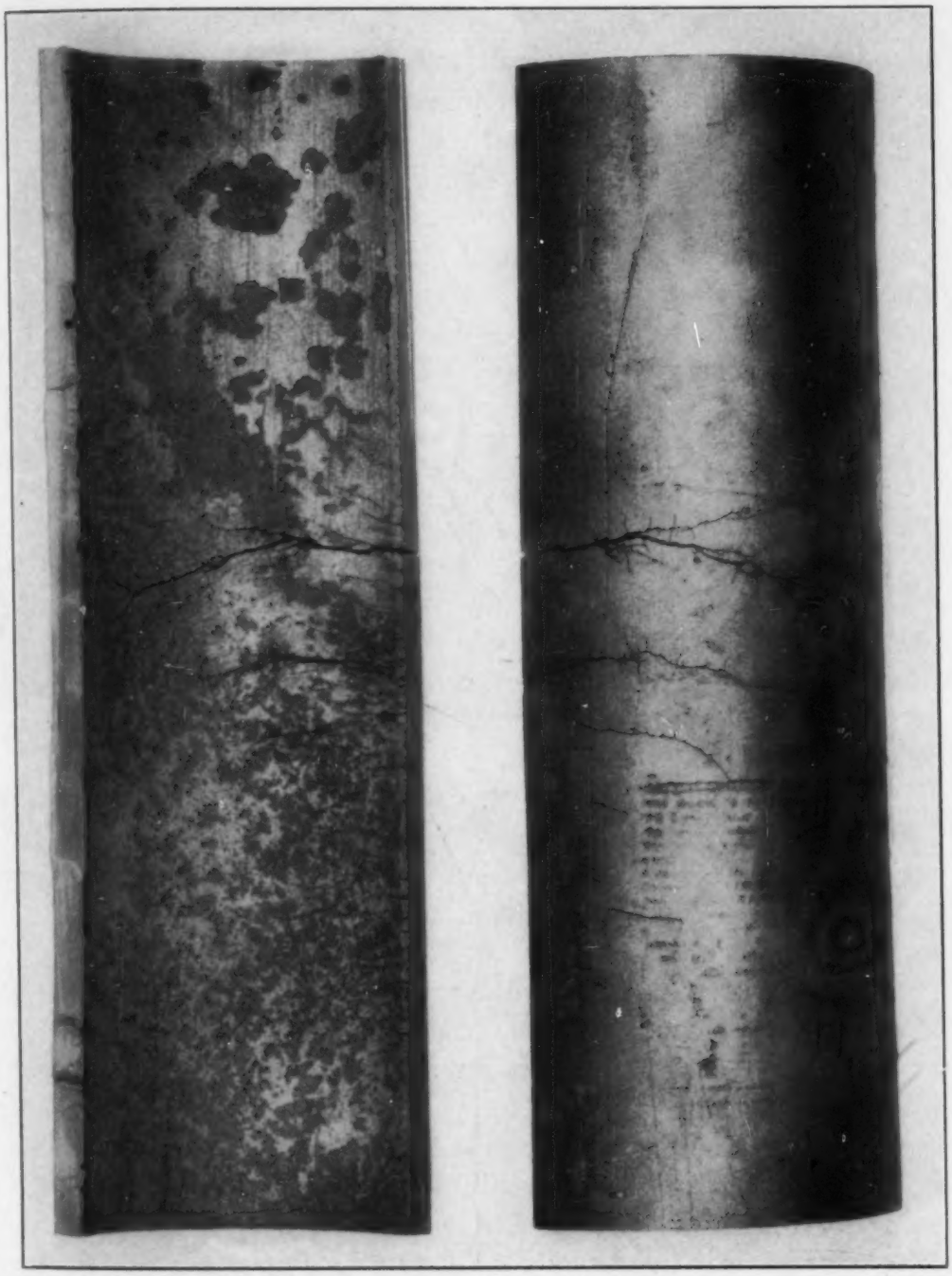

Fig. 11—Inside (Left) and Outside (Right) Views of the Central Section of Bottle "A" (18 Per Cent Chromium, 8 Per Cent Nickel) After Test. Cracking Was Both Intercrystalline and Transcrystalline.

testing of various test bottles exposed to the action of ethyl chloride plus water.

A view of one of the test bottles is presented in Fig. 10. A small hole was drilled through one of the ends and the bottles almost filled with the test solution which consisted of ethyl chloride to

which had been added 0.5 per cent of water. A copper connection tube was then brazed to the drilled end and the copper tube attached to a release valve and a pressure gage. This assembly was then placed in a water bath, and the water bath maintained at boiling temperature. As decomposition of the ethyl chloride proceeded, the internal pressure was allowed to increase to 1000 pounds per square inch equivalent to an actual maximum fiber stress of 5000 to 6000 pounds per square inch in the test bottles. When the pressure reached this value it was lowered to 400 pounds per square inch and the system again closed. It was usually necessary to so release the pressure four to five times per month of test.

*Bottle A:* (18-8 Tube Locally Torch-Heated and 18-8 Ends Welded). A 16-inch length of 2½-inch O.D. by $\frac{3}{16}$-inch wall 18-8 tubing was torch-heated to 1650 degrees Fahr. (900 degrees Cent.) on one side, the other side being retained cool by immersion in water. 18-8 cupped ends were then metal-arc welded on using 18-8 covered electrodes, and the bottle subjected to test as described. This test specimen contained severe localized stresses as a result of the torch heating and the welding. It was, of course, also susceptible to intercrystalline corrosion.

The tube failed during test, failure occurring throughout the body by intercrystalline and transcrystalline cracking. Inside and outside views of the central section of this specimen are shown in Fig. 11. Several samples cut from this tube were examined microscopically.

In the weld metal and in the weld heat affected tube material the cracking was entirely transcrystalline. Between the two welds, the cracks usually began in an intercrystalline manner. Many of these intercrystalline cracks, after progressing for a short distance, changed into transcrystalline cracks. Sometimes a transcrystalline and an intercrystalline crack would progress parallel to one another and in close proximity. Fig. 12A is a photomicrograph at 200 diameters magnification showing an intercrystalline crack at its inception at the inside surface and Fig. 12B shows the progress of this crack through the tube wall. Fig. 12C is a view of an intercrystalline crack (top of view) transforming into a transcrystalline crack during its progression through the tube wall. Fig. 12D is even more interesting in that it shows, separated by an actual distance of about 1/100 inch, two cracks, one transcrystalline (left) and the other intercrystalline.

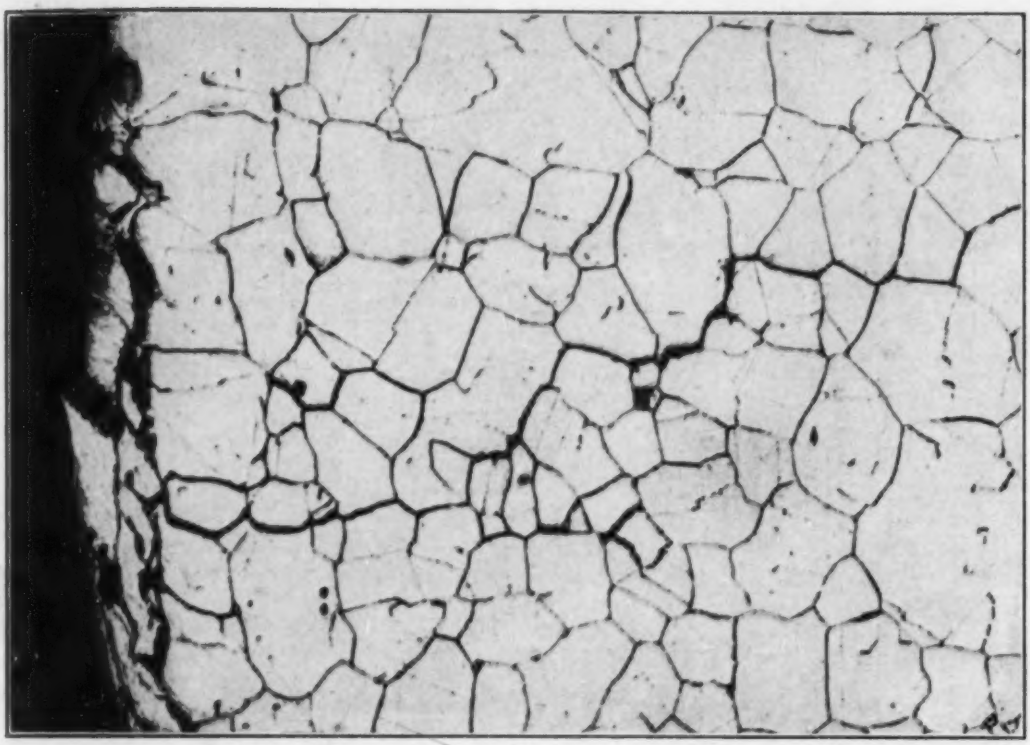

Fig. 12A—Intercrystalline Crack in Bottle "A", Fig. 11, After Testing in Ethyl Chloride Plus Water at Point of Inception at Inside Surface. × 200.

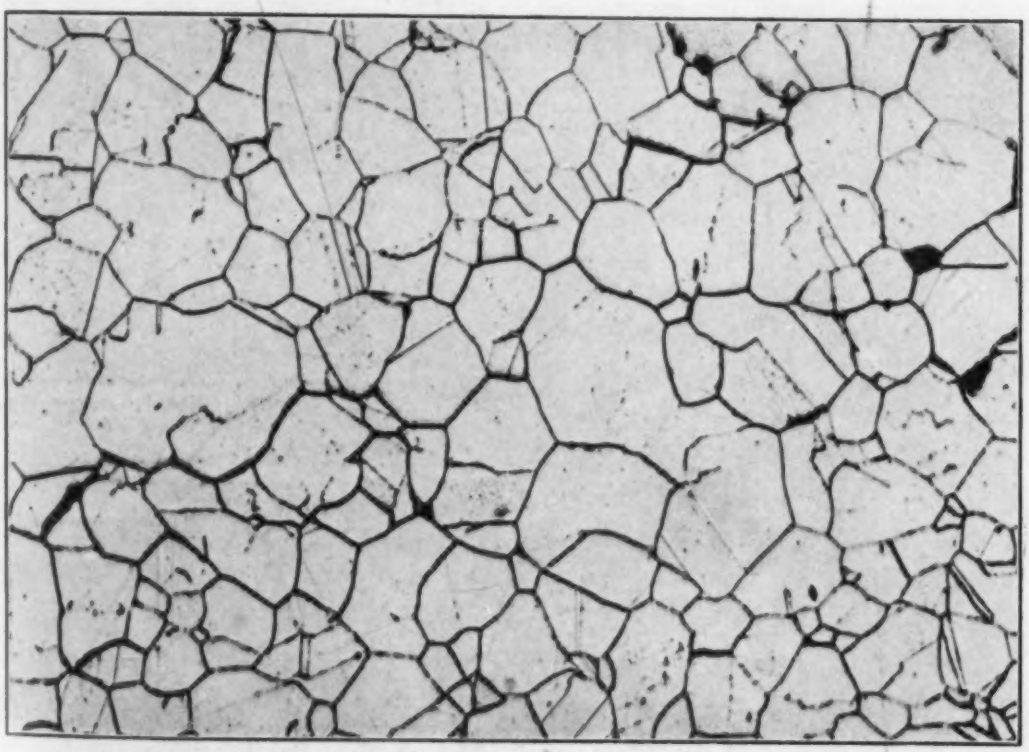

Fig. 12B—Progression of this Crack, Fig. 12A, Through the Tube Wall. × 200.

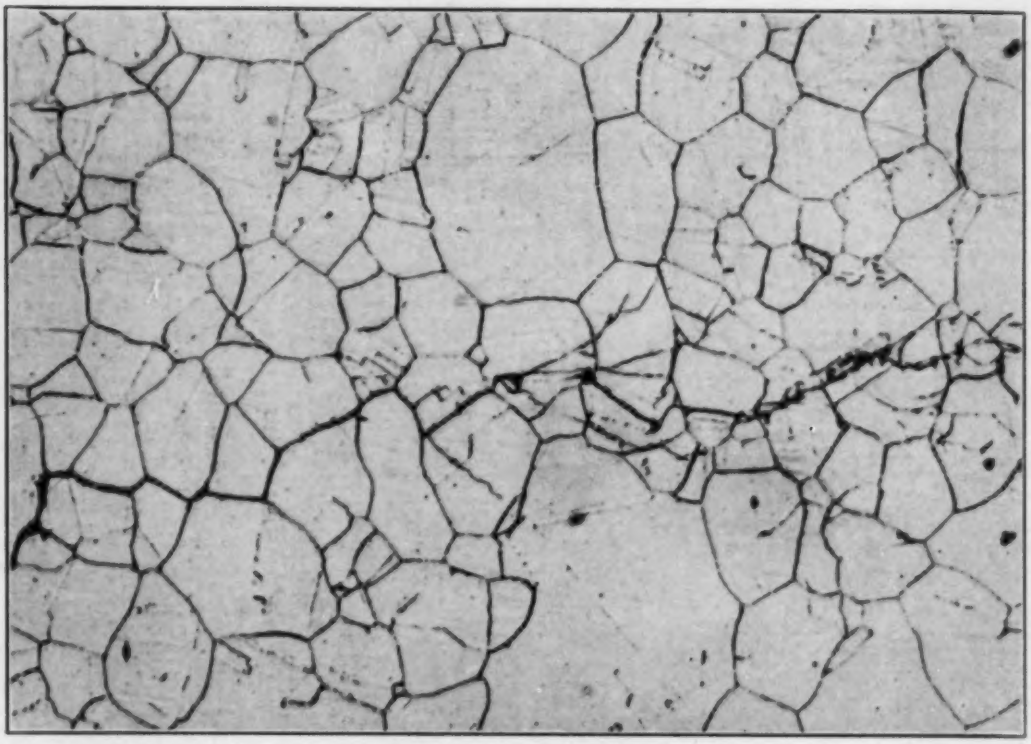

Fig. 12C—An Intercrystalline Crack Changing to a Transcrystalline Crack During Progression Through the Wall. × 200.

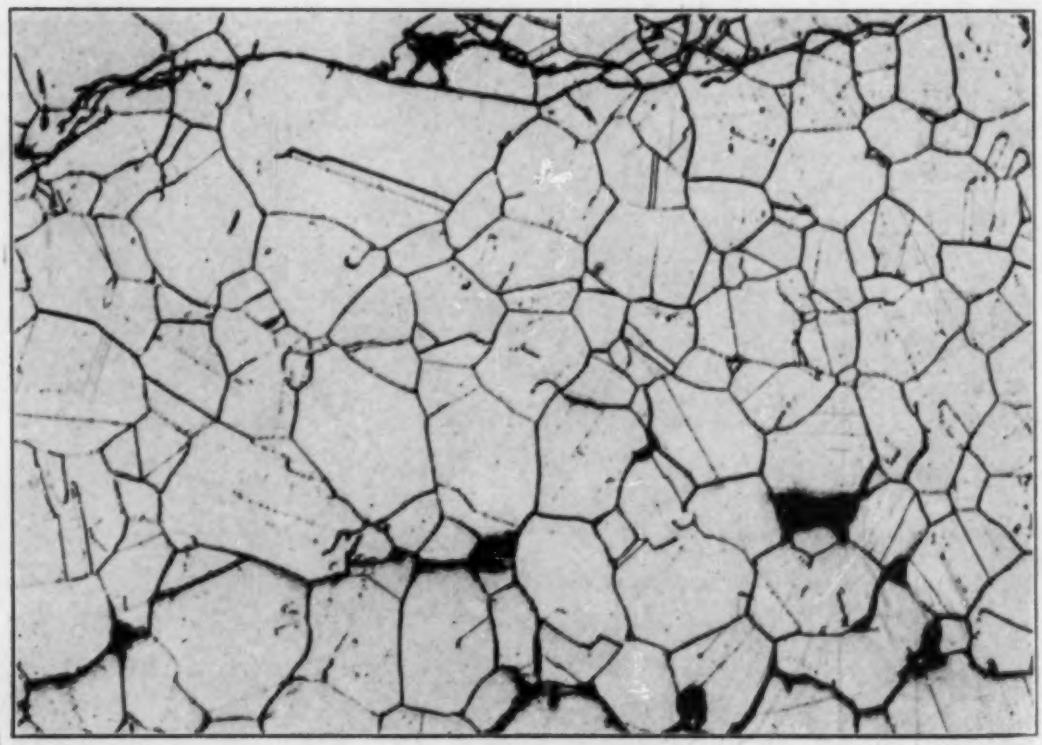

Fig. 12D—A Transcrystalline and an Intercrystalline Crack About 1/100 Inch Apart. × 200.

*Bottle B:* (18-8 Tube Stress Relieved at 1350 degrees Fahr. (730 degrees Cent.) and 18-8 Ends Welded). A 16-inch length of 2½-inch O.D. by $\frac{3}{16}$-inch wall 18-8 tubing was heated at 1350 degrees Fahr. (730 degrees Cent.) for one hour and furnace cooled. 18-8 cupped ends were then welded on both ends and the bottle tested in the as-welded condition. The bottle failed by transcrystalline cracking through the weld and the weld heat affected portions adjacent to the weld. No cracking occurred at other locations. These uncracked locations are in a condition of susceptibility to intercrystalline corrosion because of the carbide precipitation which occurred as a result of the heat treatment at 1350 degrees Fahr. (730 degrees Cent.). These areas are, however, practically free of internal stress, and stress corrosion cracking did not occur. The residual welding stresses present at and adjoining the welds have been the deciding factor in determining the location of the failure. Fig. 13A shows a transcrystalline crack at point of inception on the inside surface of tube at a location in the heat affected zone of the tube immediately adjoining the weld metal. Another view of transcrystalline cracks during progression through the wall of the tube in the heat affected region adjoining the weld metal is presented in Fig. 13B and in Fig. 13C is shown a transcrystalline crack crossing from the heat affected region into the weld metal.

*Bottle C:* (18-8 Tube with 18-8 Ends Welded and Stress Relieved at 1350 Degrees Fahr.). A 16-inch length of 2½-inch O.D. by $\frac{3}{16}$-inch wall 18-8 tubing was closed at both ends by welding on 18-8 cupped ends. The bottle was then heat treated at 1350 degrees Fahr. (730 degrees Cent.) for one hour and furnace cooled to room temperature. It was then tested in the ethyl chloride-water mixture for 1560 hours after which time the test was discontinued. Careful examination of the sectioned bottle revealed no indication of intercrystalline or transcrystalline cracking (Fig. 14). The stress relieving treatment has evidently prevented failure.

*Bottle D:* (18-8 Tube Welded to Carbon Steel Tube; 18-8 Ends Welded; and Assembly Stress Relieved at 1350 Degrees Fahr., Air Cooling from 600 Degrees Fahr.). An 8-inch length of 2½-inch O.D. by $\frac{3}{16}$-inch 18-8 steel tubing was flash welded to an 8-inch length of 2½-inch O.D. by $\frac{3}{16}$-inch wall carbon steel tubing. 18-8 cupped ends were then metal-arc welded to the two ends. The bottle so formed was heat treated at 1350 degrees Fahr. for one hour, furnace cooled to 600 degrees Fahr. and then air-cooled. Test-

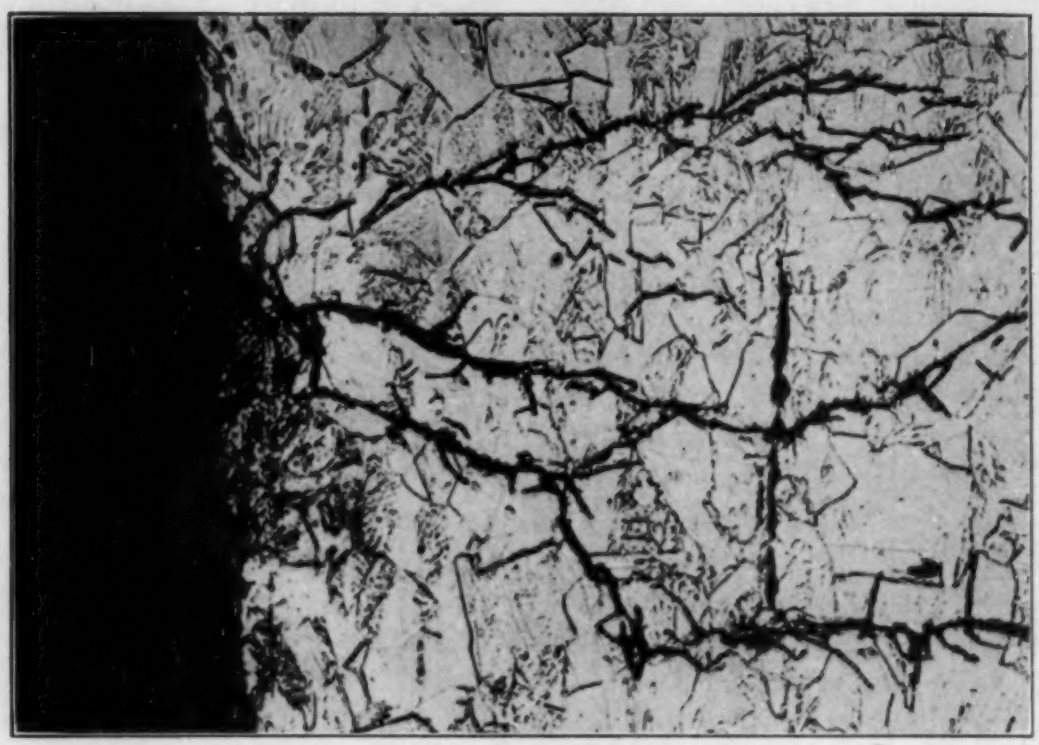

Fig. 13A—Photomicrographs of Transcrystalline Cracks Present in Bottle "B" (18 Per Cent Chromium, 8 Per Cent Nickel). Crack Beginning at Inside Tube Surface in Weld Heat Affected Region. × 200.

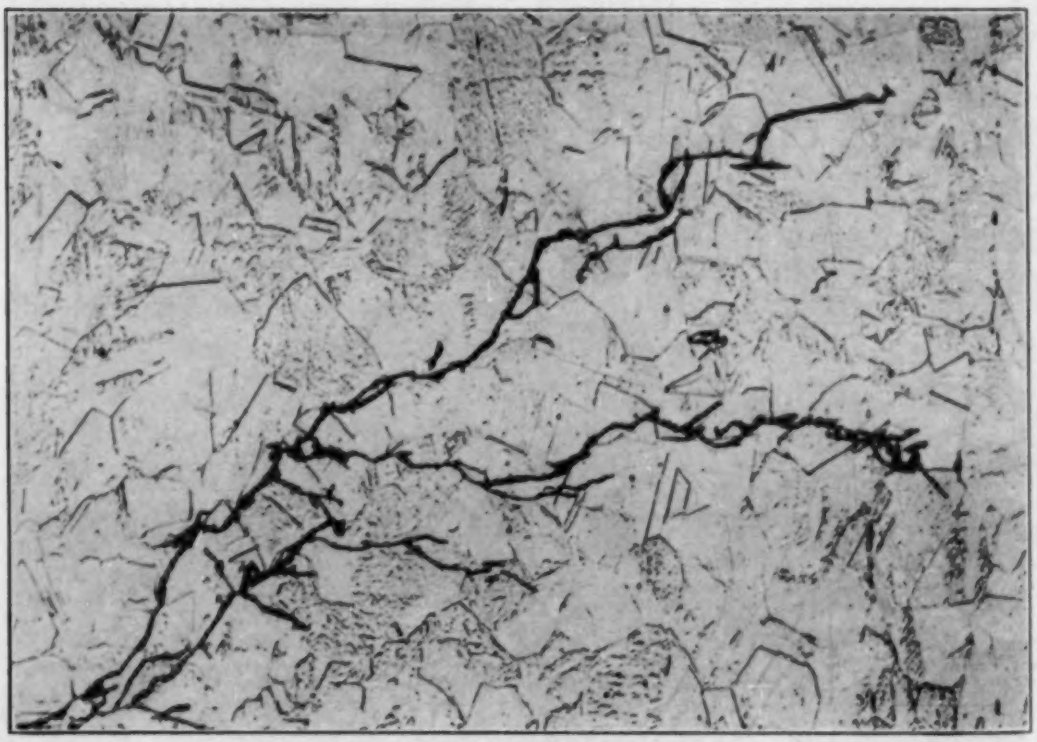

Fig. 13B—Cracks Progressing Through Weld Heat Affected Region. × 200.

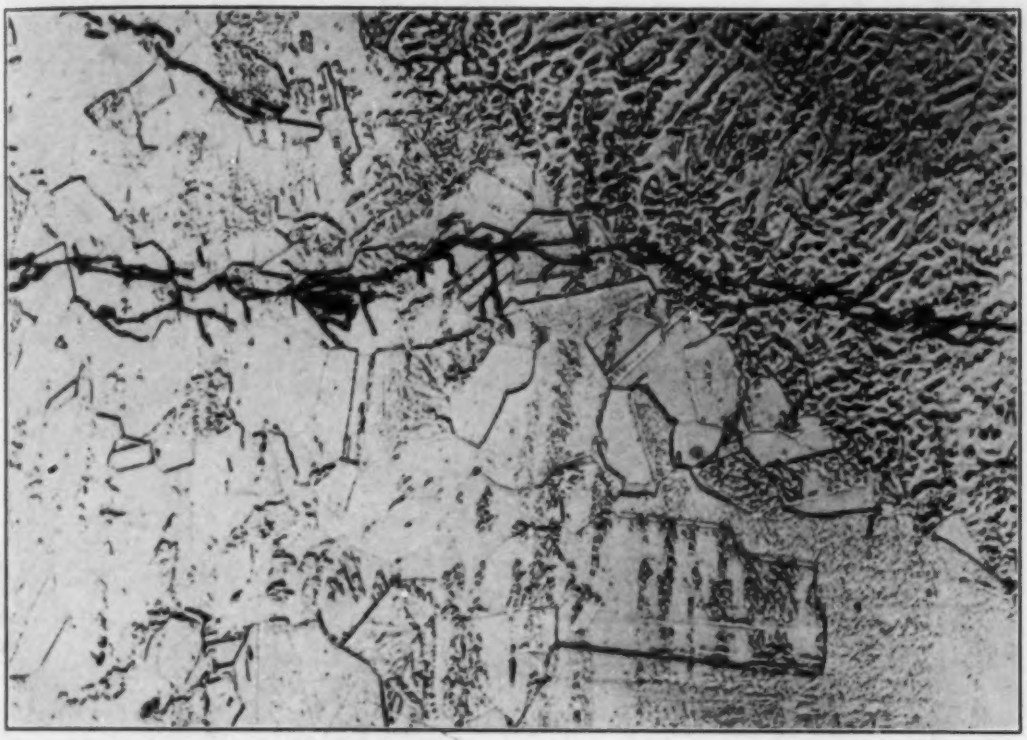

Fig. 13C—A Crack Crossing from the Weld Heat Affected Region into the Weld Metal. × 200.

ing in the ethyl chloride-water mixture developed transcrystalline cracking in the 18-8 at and adjoining the arc and the flash welds. The carbon steel was unaffected (Fig. 14). In this specimen the internal stresses, acquired by the 18-8 in the region of the welds and produced during cooling from stress relieving temperature as a result of the differences in coefficients of expansion between 18-8 and carbon steel, were sufficient to produce cracking of the 18-8. A panoramic view of one of these transcrystalline cracks is shown in Fig. 15.

*Bottle E:* (18-8 Tube Welded to Carbon Steel Tube; 18-8 Ends Welded; and Assembly Stress Relieved at 1650 Degrees Fahr., Air Cooling from 600 Degrees Fahr.). Six-inch lengths of 2½-inch O.D. by $\frac{3}{16}$-inch wall 18-8 and carbon steel tubing were flash welded together. 18-8 cupped ends were arc-welded on and the bottle so formed was annealed at 1650 degrees Fahr. (900 degrees Cent.) for one hour, furnace-cooled to 600 degrees Fahr. and then air-cooled. Examination after 1560 hours of test showed the presence of several fine cracks in the 18-8 adjoining the flash weld.

These cracks were both intercrystalline and transcrystalline. The

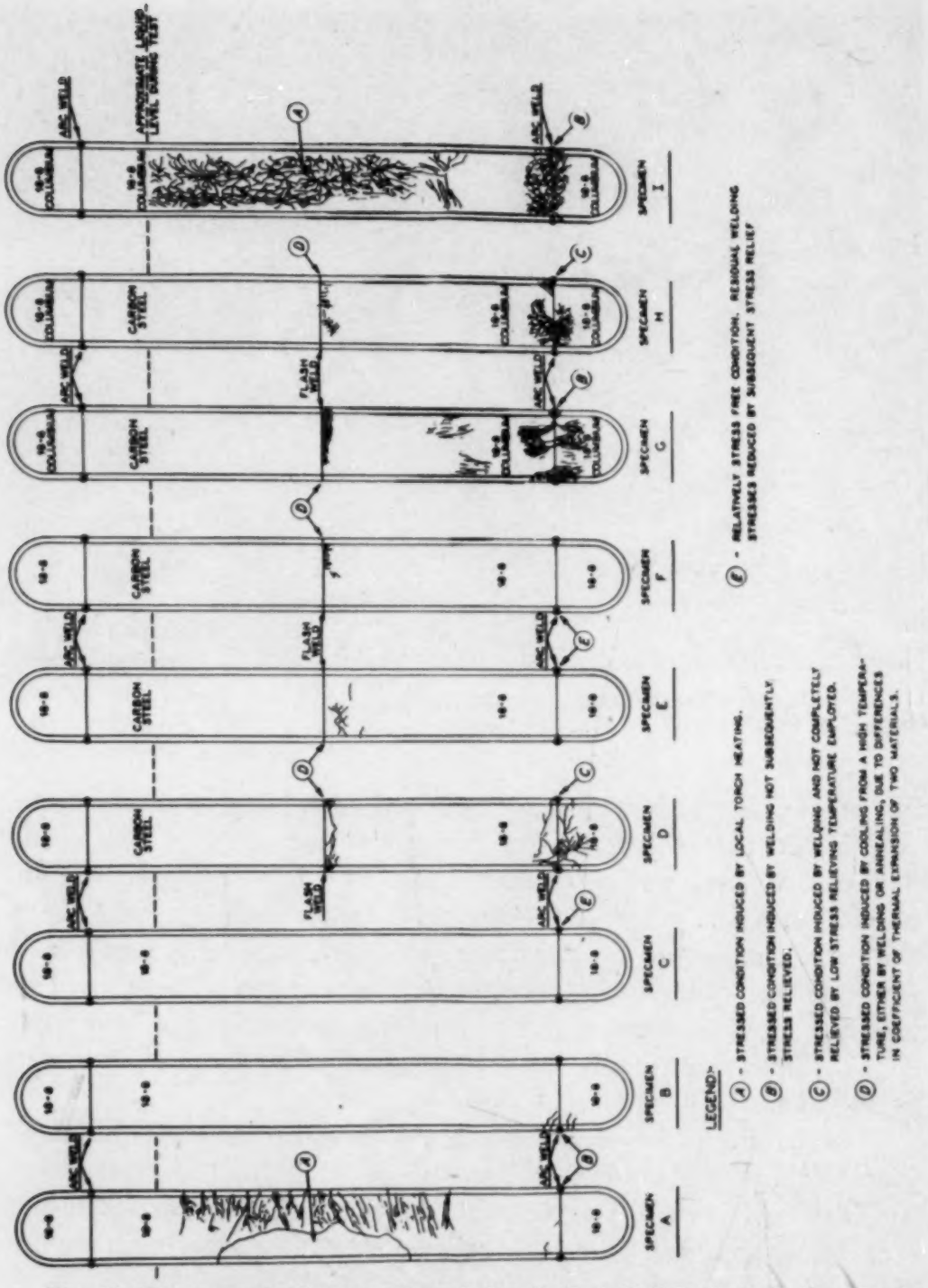
SPECIMEN I
SPECIMEN H
SPECIMEN G
SPECIMEN F
SPECIMEN E
SPECIMEN D
SPECIMEN C
SPECIMEN B
SPECIMEN A
CARBON STEEL
ARC WELD
FLASH WELD
LEGEND-
(A) - STRESSED CONDITION INDUCED BY LOCAL TORCH HEATING.
(B) - STRESSED CONDITION INDUCED BY WELDING NOT SUBSEQUENTLY STRESS RELIEVED.
(C) - STRESSED CONDITION INDUCED BY WELDING AND NOT COMPLETELY RELIEVED BY LOW STRESS RELIEVING TEMPERATURE EMPLOYED.
(D) - STRESSED CONDITION INDUCED BY COOLING FROM A HIGH TEMPERATURE, EITHER BY WELDING OR ANNEALING, DUE TO DIFFERENCES IN COEFFICIENT OF THERMAL EXPANSION OF TWO MATERIALS.
(E) - RELATIVELY STRESS FREE CONDITION. RESIDUAL WELDING STRESSES REDUCED BY SUBSEQUENT STRESS RELIEF

Fig. 14—Diagrammatic Representation of Longitudinal Half-Sections of Test Bottles "A" to "I", Inclusive, After Testing in the Ethyl Chloride-Water Mixture. The Locations and Approximate Number of the Stress Corrosion Cracks Present are Indicated.

carbon steel was unaffected (Fig. 14). This sample differed from Bottle D in that a much higher stress relieving temperature was employed in order to insure complete removal of all welding

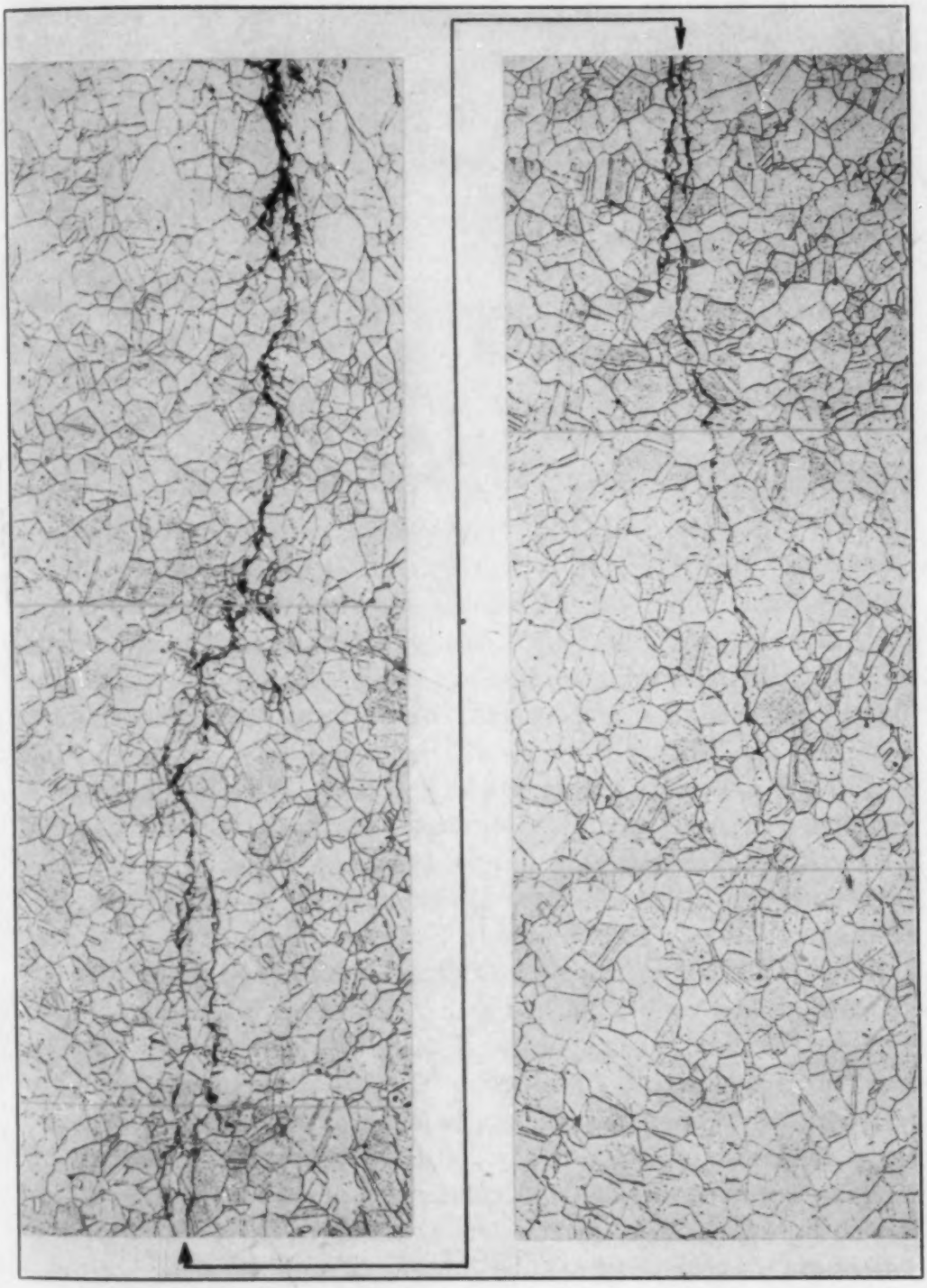

**Fig. 15—Photomicrographs Showing a Panoramic View of One of the Transcrystalline Cracks Present in Bottle "D", Fig. 14. × 200.**

stresses. The results secured furnish more conclusive proof of the presence of dangerous stresses in 18-8 produced as a result of the differing expansion coefficients of the two metals.

*Bottle F:* (18-8 Tube Welded to Carbon Steel Tube; 18-8 Ends Welded; and Assembly Stress Relieved at 1650 Degrees Fahr., Furnace Cooling to Room Temperature). This bottle was constructed of the same materials and in a manner identical with that of Bottle E. The heat treatment differed only to the extent that after annealing at 1650 degrees Fahr. the sample was furnace cooled to room temperature. After 1560 hours of test the bottle was examined. A few very fine cracks located just outside of the flash weld junction were found. These cracks were about 1/100 inch deep and originated as intercrystalline fissures but changed to transcrystalline cracks while progressing. The carbon steel was unaffected (Fig. 14). Furnace cooling to room temperature reduced the amount of stressing, but did not completely prevent it and eventual failure would have occurred.

*Bottle G:* (18-8 Cb. Tube Welded to Carbon Steel Tube; 18-8 Cb. Ends Welded). A 6-inch length of 2½-inch O.D. by $\frac{3}{16}$-inch wall 18-8 columbium-bearing tubing stabilized by heat treatment at 1600 degrees Fahr. for four hours was flash welded to a length of carbon steel tubing of the same dimensions. 18-8 columbium-bearing cupped ends were arc-welded on and the bottle tested in the as-welded condition.

Rapid failure by transcrystalline cracking occurred in the 18-8 columbium sections at and adjoining the arc and flash welds. The carbon steel was unaffected (Fig. 14).

*Bottle H:* (18-8 Cb. Tube Welded to Carbon Steel Tube; 18-8 Cb. Ends Welded; and Assembly Stress Relieved at 1350 Degrees Fahr., Air Cooling from 600 Degrees Fahr.). A bottle was constructed of the same materials and in a manner identical with that of G. The sample was annealed at 1350 degrees Fahr. for one hour, furnace-cooled to 600 degrees Fahr. and then air-cooled. This heat treatment duplicated that given Bottle D which was constructed using 18-8 not columbium stabilized. Failure by transcrystalline cracking occurred in the 18-8 columbium at and adjoining the arc and flash welds. The carbon steel was unaffected (Fig. 14).

*Bottle I:* (18-8 Cb. Tube Locally Torch Heated; 18-8 Cb. Ends Welded). A 12-inch length of 2½-inch O.D. by $\frac{3}{16}$-inch wall 18-8 columbium stabilized tubing was locally torch-heated on one side to a temperature of approximately 1650 degrees Fahr. (900 degrees Cent.), the other side being maintained cool by immersion in water. 18-8 columbium cupped ends were arc-welded

Fig. 16—View Showing the Cracked Condition After Test of Sections of Bottle "I" (18 Per Cent Chromium, 8 Per Cent Nickel Plus Columbium) Which Had Been Locally Torch Heated. All Cracks are Transcrystalline.

on and the bottle tested in this condition. This sample is identical with Bottle A except for the columbium content of Bottle I. Rapid transcrystalline failure occurred at the arc welds and throughout the body of the bottle where it had been torch-heated (Fig. 14). Fig. 16 shows the cracked condition of the central portion of this sample. The pattern of stress corrosion cracking produced is of interest, indicating many centers from which the cracks developed and indicating the random distribution and direction of the internal stresses.

It is evident that the 18-8 columbium material is at least as susceptible to stress corrosion cracking under the conditions of this test as is ordinary 18-8. It is remarkable that the carbon steel in all of the ethyl chloride-water tests has remained totally unaffected whether in a highly stressed or in a stress-free condition. The conclusion is inevitable that austenitic chromium-nickel steels, whether susceptible to intercrystalline corrosion or not, are highly susceptible to stress corrosion failure under conditions of internal stress and exposure to certain corrosive media ordinarily considered only mildly corrosive.

This ethyl chloride-water mixture offers a remarkably definite means of detecting stress in the chromium-nickel austenitic steels. It is possible that other face-centered cubic metals or alloys (copper, brass, bronze, nickel, etc.) may respond in a similar manner. No

ethyl chloride experiments on such metals have as yet been attempted.

The conclusions that may be drawn from the results of these ethyl chloride tests may be summarized as follows:

1. An austenitic stainless steel susceptible to intercrystalline corrosion will not develop intercrystalline or transcrystalline cracking when in contact with a corrosive medium provided that the steel is free of internal stress.
2. An austenitic stainless steel susceptible to intercrystalline corrosion may develop either transcrystalline or intercrystalline cracking (or both) when in contact with a corrosive medium if the steel is not free of internal stresses.
3. Intercrystalline cracking induced by the combination of stress and corrosion may be differentiated from intercrystalline attack caused by a corrosive medium only. When stresses are present the attack will be very localized, whereas when stresses are not present the intercrystalline attack will be general.
4. Austenitic stainless steels free from intercrystalline corrosion by reason of heat treatment or by the use of stabilizing elements but containing internal stresses may fail rapidly by transcrystalline cracking when in contact with a corrosive medium. The same steel free of stresses will not develop cracking irrespective of inherent corrosion resistance.
5. A stressed condition of dangerous magnitude may be readily induced by heat treatments involving rapid cooling of the austenitic steels. Such rapid cooling treatment could leave a stress-free condition only if all portions of a part or structure could be cooled simultaneously and at the same rate. This is a practical impossibility.
6. When austenitic steels of relatively high coefficient of thermal expansion are joined to ferritic steels, as by welding, an internally stressed condition is established, which cannot be completely eliminated by subsequent heat treatment, even with extremely slow rates of cooling. The magnitude of these stresses, induced in the austenitic steel in such composite structures, is sufficiently high to produce stress corrosion cracking when exposed to selected media, such as water-bearing ethyl chloride.

### Effect of Various Corroding Media and Magnitude of Applied Tensional Stress

The experiments, in which cracking was produced in stressed

austenitic alloy steels by a mixture of ethyl chloride and water, satisfactorily explained the service failures of the thermowells. No direct relationship, however, existed between the experimental stress corrosion cracking and the service failures of the 18-8 superheater tubes referred to earlier. While the ethyl chloride experiments confirmed

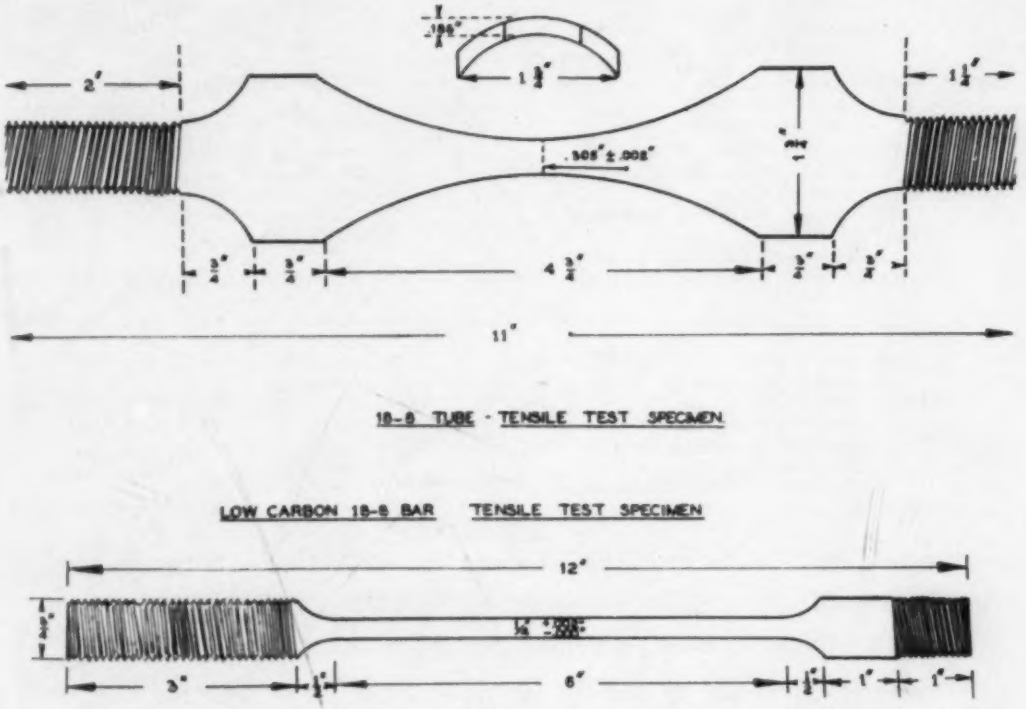


Fig. 17A—Sketch Showing the Type of Specimen Machined From 18 Per Cent Chromium, 8 Per Cent Nickel Tubing and Used for the Tensile Stress Corrosion Experiments.

Fig. 17B—Sketch Showing Type of 18-8 Bar Specimen Used for the Tensile Stress Corrosion Experiments.

the nature and mechanism of the superheater failures, they did not directly establish the active corrosion medium responsible for the corrosion factor in the service failures. The superheater elements had been subjected to hydrostatic test water and in service to relatively pure water or steam.

Attempts were made to experimentally reproduce stress corrosion cracking in the 18-8 superheater tubing using stock used for the original installation. These 2½-inch O.D. by $\frac{3}{16}$-inch wall tubes were flash resistance welded and heat treated at 1350 degrees Fahr. (730 degrees Cent.) for one hour in order to secure the same struc-

tural condition as was present in the original. Samples were then stressed by various means such as by bending or by local heating and subjected to (a) hydrostatic test water, (b) 1 per cent by weight hydrochloric acid solution and (c) 1 per cent by weight sulphuric acid solution and (d) moist atmospheric conditions for time periods up to one year. No cracking was produced. These test results indicated that the combination of conditions responsible for the serv-

**Table II**

**18-8 Tube Specimens**

**Tested for 840 Hours Under a Constant Tensile Load of 35,000 Pounds Per Square Inch**

| Specimen | Corrosive Agent | Results of Test |
|---|---|---|
| 1 | Hydrostatic Test Water | No cracking |
| 2 | 1% Hydrochloric Acid (by weight) in Distilled Water | No cracking |
| 3 | 5% Sodium Chloride Solution | No cracking |
| 4 | 6% Sodium Chloride 3% Hydrogen Peroxide Solution | No cracking |
| 5 | 6.6% Ferric Chloride Solution | Intercrystalline cracking produced failure in 4 days |
| 6 | Saturated Solution of Mercuric Chloride | Intercrystalline cracking produced failing in 2 days |
| 7 | 5% Magnesium Chloride Solution | No cracking |
| 8 | Saturated Solution of Sodium Nitrate | No cracking |
| 9 | 1% Sulphuric Acid (by weight) in Distilled Water | Severe intercrystalline cracking but complete failure did not occur |
| 10 | 5% Sodium Sulphate Solution | No cracking |
| 11 | 25% Sodium Hydroxide 3% Sodium Silicate Solution | No cracking |

ice failure had not been approximated and that stress corrosion cracking would be developed only by specific corrosive media.

In an effort to evaluate quantitatively the amount of stress required to produce stress corrosion failure and to isolate from a selected list the corrosive media capable of producing cracking under the proper stress conditions, the following experiments were conducted.

A. Tensile specimens having a variable cross-sectional area were machined from 2½-inch O.D. by $\frac{3}{16}$-inch wall 18-8 tubing. The type of specimen used is indicated in Fig. 17A. The machined specimens were heated at 1350 degrees Fahr. for one hour and slowly furnace cooled. This treatment was imposed to eliminate stresses previously present and to insure susceptibility to intercrystalline corrosion. Chemical analysis of all specimens was as follows:

| Carbon | Manganese | Silicon | Chromium | Nickel |
|---|---|---|---|---|
| 0.07 | 0.38 | 0.23 | 17.37 | 9.22 |

By means of springs a tensile stress of 35,000 pounds per square inch for the smallest cross-sectional area was imparted

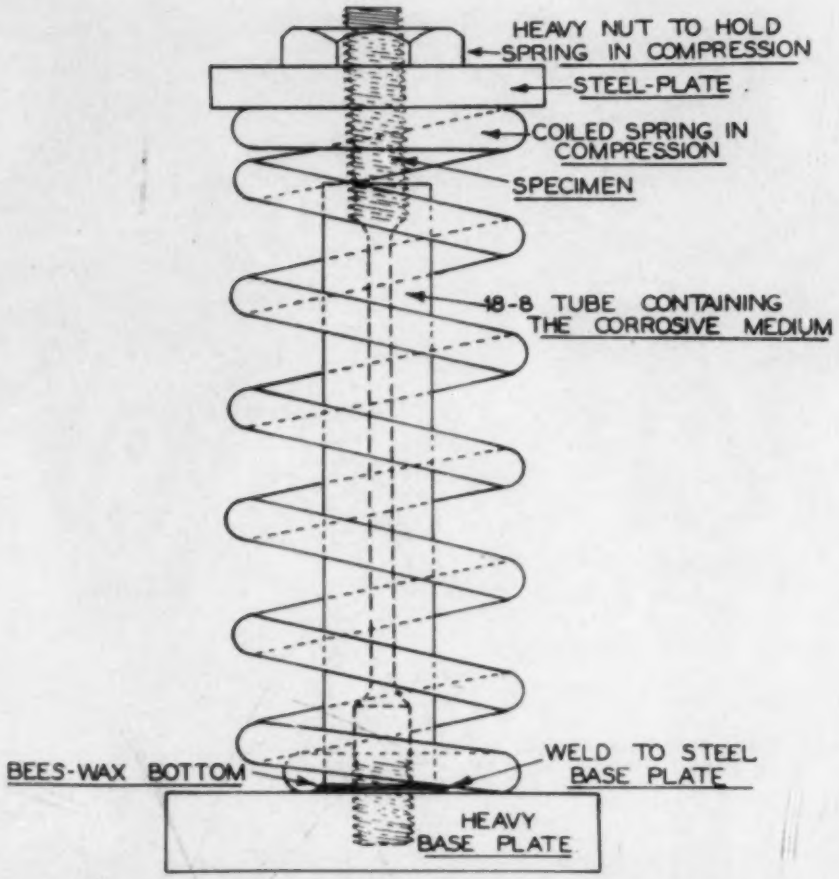


Fig. 18—Diagrammatic Sketch Indicating the Type of Apparatus Used in the Tensile Stress Corrosion Experiments.

for a period of 840 hours. The test equipment used was essentially as indicated by Fig. 18, the specimens being under constant tension while immersed in the corrosive media. The corroding agents used and the results of the test are summarized in Table II.

B. In view of the effectiveness of ferric chloride in producing intercrystalline cracking, the next endeavor was to determine whether low tensile stresses, and low ferric chloride concentrations would still result in cracking. Tube specimens duplicating the material, shape and heat treatment of those described under A were tested under tensile loads of 35,000, 20,000 and 10,000 pounds per square inch while immersed in 1.25, 0.25 and 0.05 per cent ferric chloride solutions. The test was conducted for a total period of 1600 hours, during which time two of the specimens fractured. Complete test data are given in Table III. This table indicates that a defi-

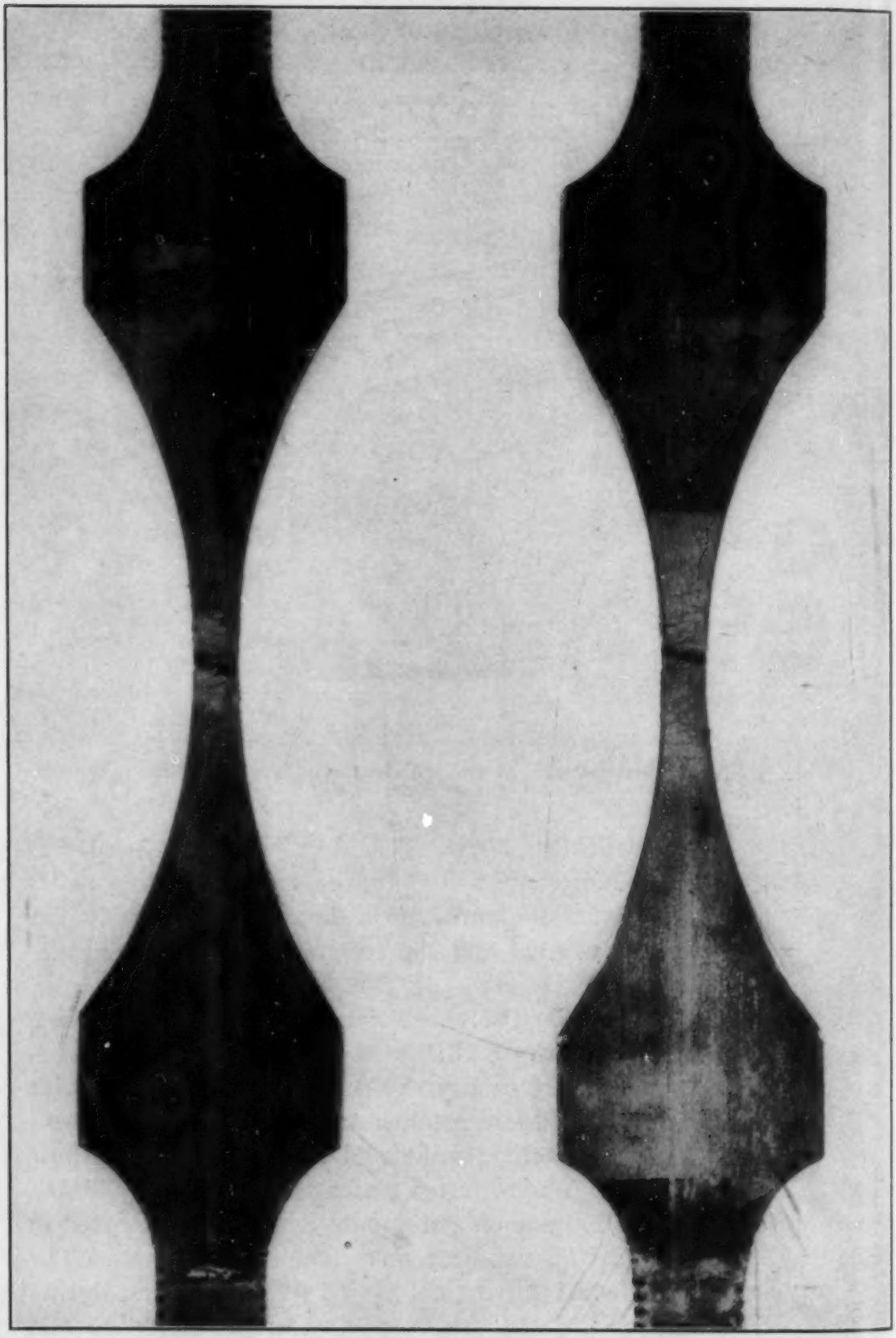

Fig. 19—Views of Two of the 18-8 Tube Specimens (Specimens 3 and 6 of Table III) Showing Manner of Failure Under Tensile Stress. All Cracks are Intercrystalline at Point of Inception.

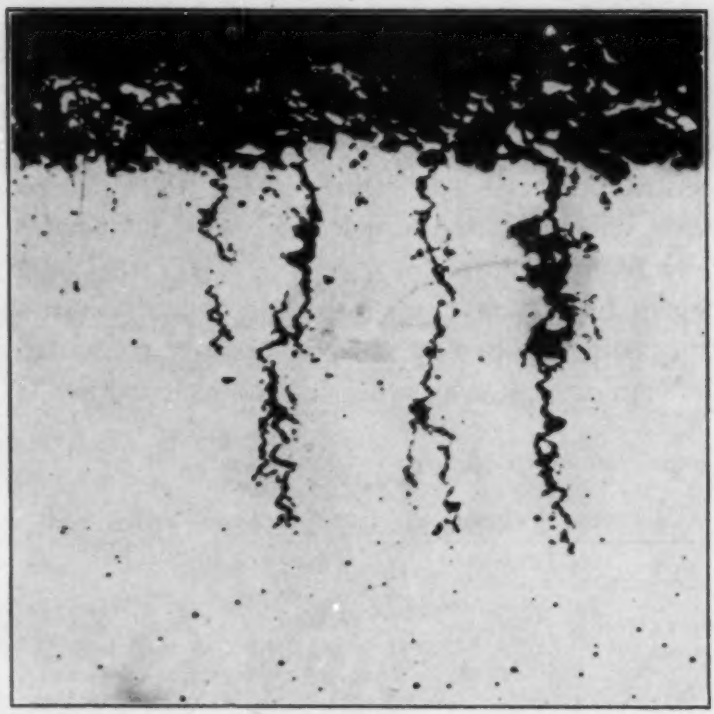

Fig. 20A—Unetched Specimen of Intercrystalline Cracks Representative of Those Present in Specimens of Table IV. × 50.

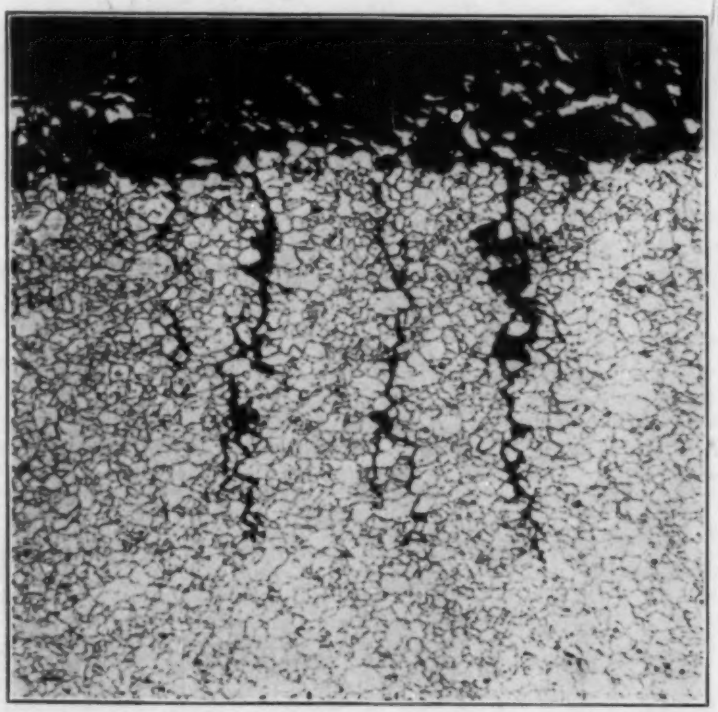

Fig. 20B—Etched Specimen of Intercrystalline Cracks Representative of Those Present in Specimens of Table IV. × 50.

nite correlation exists between stress intensity and degree of cracking and between ferric chloride concentration and cracking as may be noted from the rapidity of failure column in Table III. From the results secured, it is indicated that concentrations well below the value of 0.05 per cent of ferric chloride will ultimately crack a stressed material. All cracks were intercrystalline. Views of two failed specimens are shown in Fig. 19 and unetched and etched views showing the intercrystalline nature of the cracking are presented in Fig. 20.

C. To determine if it was possible to develop cracking in 18 per

**Table III**

**18-8 Tube Specimens—$\frac{5}{16}$ Inch Thick—1600 Hours Test**

| Specimen | Concentration of Ferric Chloride Per Cent | Constant Tensile Stress Applied Lbs. Per Sq. In. | Remarks on Condition of Samples After Test | Rating in Order of Rapidity With Which Failure Was Occurring |
|---|---|---|---|---|
| 1 | 1.25 | 10,000 | Severe intercrystalline cracking at minimum cross-section | 4 |
| 2 | 1.25 | 20,000 | Same as above | 3 |
| 3 | 1.25 | 35,000 | Failed through minimum cross-section after 840 hours | 1 |
| 4 | 0.25 | 10,000 | Severe intercrystalline cracking at minimum cross-section | 6 |
| 5 | 0.25 | 20,000 | Same as above | 5 |
| 6 | 0.25 | 35,000 | Failed through minimum cross-section after 980 hours | 2 |
| 7 | 0.05 | 10,000 | Moderate intercrystalline cracking at minimum cross-section. $\frac{1}{8}$ through wall | 9 |
| 8 | 0.05 | 20,000 | Same as above, $\frac{1}{4}$ through wall | 8 |
| 9 | 0.05 | 35,000 | Same as above, $\frac{1}{2}$ through wall | 7 |

**Table IV**

**Low Carbon 18-8 Bar Specimens—$\frac{1}{4}$ Inch Diameter—1100 Hour Test**

| Number | Concentration of Ferric Chloride Per Cent | Constant Tensile Stress Applied Pounds Per Sq. In. | Remarks on Condition of Specimens After Test |
|---|---|---|---|
| 1 | 1.25 | 10,000 | Very slight general intercrystalline corrosion. A few large and deep pits. No cracking. |
| 2 | 1.25 | 20,000 | Very slight general intercrystalline corrosion. Several large and deep pits. No cracking. |
| 3 | 1.25 | 35,000 | Very slight general intercrystalline corrosion. Several shallow cracks. Few large and deep pits. |
| 4 | 0.25 | 10,000 | Slight general intercrystalline corrosion. Many large and deep pits. No cracking. |
| 5 | 0.25 | 20,000 | Slight general intercrystalline corrosion. A few large and deep pits. No cracking. |
| 6 | 0.25 | 35,000 | One shallow intercrystalline crack. No pits. |
| 7 | 0.05 | 10,000 | Many small pits. No cracking. |
| 8 | 0.05 | 20,000 | A few small pits. No cracking. |
| 9 | 0.05 | 35,000 | A few very shallow intercrystalline cracks. No pits. |

cent chromium-8 per cent nickel alloy of very low carbon content some 18-8 bars of the following analysis were secured and tested in a manner identical with that employed for the tube specimens described under B.

| Carbon | Manganese | Silicon | Chromium | Nickel |
|---|---|---|---|---|
| 0.050 | 0.40 | 0.27 | 18.54 | 8.38 |

The reduced section of these machined tensile specimens had a constant cross-sectional area as shown in Fig. 17B. The rod was water-quenched from 1950 degrees Fahr. (1065 degrees Cent.) and then heated at 1350 degrees Fahr. (730 degrees Cent.) for one hour, followed by furnace cooling. The copper sulphate test indicated a slight, but definite, susceptibility to intercrystalline corrosion. The test was conducted for a period of 1100 hours using the same stresses and the same ferric chloride concentrations as in B. No failure occurred during the period of test. Complete test data are given in Table IV. All but one of the specimens developed rather deep pits, but only four showed the presence of a few shallow cracks. The pitting was more serious than the intercrystalline cracking because it was much deeper. These samples are but very slightly susceptible to intercrystalline corrosion and serious cracking did not develop. The test results indicate that ferric chloride causes accelerated pitting when it cannot produce intercrystalline cracking. The results indicate that a columbium or titanium stabilized alloy would be immune to intercrystalline corrosion under the conditions of this test, as would be a water-quenched 18-8 without columbium or titanium.

## General Discussion

A number of service failures of the austenitic chromium-nickel steels have been encountered, these failures having been identified as due to stress corrosion cracking. These cracks were intercrystalline in some cases and transcrystalline in others, the path of the fracture being dependent upon the corrosive medium, the stress magnitude, and the degree of susceptibility to intergranular attack.

The ethyl chloride experiments revealed the extreme susceptibility of these steels to this type of failure under conditions of stress and corrosion that left carbon steel unaffected. Tests using other

corrosive agents indicated that relatively few of them are capable of producing cracking in stressed 18-8. Ferric chloride and mercuric chloride over wide concentration ranges are, however, very effective in producing intergranular stress corrosion cracking in susceptible material. Other mildly corrosive agents are also capable of producing intergranular cracking in susceptible austenitic steels under stress. For example, the failed superheater tubing had only been exposed to the action of the hydrostatic test water and to relatively pure boiler water and steam although some slight concentration of dissolved salt in the test water, such as ferric chloride, may have occurred.

The only corrosive medium used in the experimental tests which has been definitely identified as capable of producing both transcrystalline and intercrystalline cracking in stressed austenitic steels is the ethyl chloride-water mixture.

Undoubtedly other reagents in addition to the ethyl chloride-water mixture are effective in producing both transcrystalline and intercrystalline cracking. Because of the extensive and generally satisfactory use of these austenitic steels under conditions of stress, the number of such reagents is probably small although the histories of a few transcrystalline failures that have developed in oil processing[1] and in the manufacture of soap indicate that the ethyl chloride is not unique in its action. Both ferric chloride and mercuric chloride while very active in producing intercrystalline cracking in samples susceptible to intercrystalline corrosion were incapable of producing either intercrystalline or transcrystalline cracking in samples not susceptible. The inference is that material not susceptible to intercrystalline corrosion is less liable to stress corrosion failure than is susceptible material.

The minimum stress necessary for the production of cracks under various corrosive conditions has not been definitely established. The experimental evidence would indicate that stress values below the yield point value are sufficient for the production of intergranular cracks in susceptible material. The stress required to produce a transcrystalline failure in either susceptible or unsusceptible material is probably in excess of the yield point or at least in excess of the elastic limit of the metal. This has not been definitely established, however, and work now in progress includes a deter-

[1] "Transcrystalline Disintegration of 18-8 Steel," *Mining and Metallurgy*, American Institute of Mining and Metallurgical Engineers, March, 1936, p. 159.
See also "Book of Stainless Steels," 2nd Ed., p. 591, published by the American Society for Metals.

mination of the minimum stress requirement. Since stress corrosion cracking in unsusceptible material is always transcrystalline, it follows that from the stress factor such material is again less liable to failure than susceptible material.

The true elastic limit of the various austenitic chromium-nickel compositions, particularly when in the annealed (water-quenched) state is controversial, the reported values being to a considerable extent dependent upon the possible accuracy of the equipment employed in making such determinations. The information secured and summarized in Table III indicated that intercrystalline cracking may be produced in susceptible material at a stress value of 10,000 pounds per square inch, which is well below published values for the elastic limit of any of the 18-8 type alloys. There is a possibility that because of the type of specimen employed in this test local stress concentration was present at the specimen edges.

High internal stresses in excess of yield point values may be readily introduced by many fabricating operations such as cold bending, welding and the usually employed high temperature heat treatment for the retention in solution of the chromium carbides. Corrosion cracking in service has been experienced by reason of stress factors arising from each of these treatments. The remedy lies in the elimination or reduction to a safe value of any internal stress. This can only be accomplished by thermal stress relief utilizing relatively slow rates of cooling. This is borne out by the absence of damage to material made extremely susceptible to intercrystalline corrosion but in a stress-free condition by reason of thermal stress relief at a temperature within the carbide precipitation range. The elimination of the possibility of stress corrosion cracking in 18-8 not containing a stabilizing element such as titanium or columbium introduces the possibility of failure by general intercrystalline attack. For service temperatures within the carbide precipitation range (800 to 1650 degrees Fahr.) (900 degrees Cent.) the use of titanium or columbium will provide an additional factor of safety by eliminating the possibility of intercrystalline cracking which has been shown to occur more readily than does transcrystalline cracking. These stabilized alloys offer no assurance, however, that under the proper stress and corrosion conditions a transcrystalline crack will not develop. A stress-free structure also free from susceptibility to intercrystalline corrosion may be produced by the use of stabilized alloys finally annealed and slowly cooled.

In composite structures in which austenitic steel, whether or not stabilized, is rigidly attached to other metals having different coefficients of thermal expansion, stress elimination is not possible. Cooling of such composite structures from annealing temperatures regardless of how slow the cooling rate necessarily re-establishes a high internally stressed condition. When such a situation exists cracking under critical corrosive conditions occurs.

Although our studies have been limited to the austenitic steels of the 18 per cent chromium-8 per cent nickel type with and without molybdenum and with and without the stabilizing element columbium, it is probable that other types of the group of alloys behave in a like manner.

The phenomenon of stress corrosion cracking is encountered in a number of alloys, the most notable examples being the "season cracking" of brass. Other alloys possessing the face-centered cubic lattice, including nickel and duralumin, are also prone to this type of failure. It is rarely observed in ferritic steels although the caustic embrittlement of boiler plate is an essentially similar occurrence. In all of these cases the path of the fractures is intergranular. Information relating to the intergranular failure of austenitic steels has been recently published.[2]

The transcrystalline type of stress corrosion cracking of the austenitic steels described herein is, therefore, unusual.

Questions arise as to the mechanism of stress corrosion cracking and as to any differences between this phenomenon and related types of failures such as intercrystalline corrosion without stress involved and fatigue failures with or without the element of corrosion present.

Corrosion of a metal usually results from the uniform solution of the metal surface in a solvent or by a more localized reaction called pitting. Pitting may result from a concentration of the corrosive medium and the establishment of an electrolytic couple which accelerates local solution of the metal. Stress may be entirely absent but if present will usually manifest itself by increased rate of solution.

Intercrystalline corrosion in the austenitic chromium-nickel steels results from a lowered corrosion resistance at grain boundaries produced by localized withdrawal of chromium from solution.

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[2]S. L. Hoyt and M. A. Scheil, "Stress-Corrosion Cracking in Austenitic Stainless Steels," TRANSACTIONS, American Society for Metals, 1938.

Such intercrystalline corrosion is undoubtedly accelerated because of the presence of the resistant matrix and resistant chromium carbides in contact with the impoverished areas, thus establishing ideal conditions for the formation of an electrolytic couple. In the absence of stress such intercrystalline attack will occur over the entire metal surface, the grain boundaries being completely undermined allowing the grains to fall away. With high stress present in such a material the corrosion is localized to a more limited number of grain boundaries and proceeds through the body of the material in a general direction normal to that of the applied stress. While numerous paths of stress corrosion failure or cracks may be produced many of the grain boundaries will remain unattacked. Moreover, in such material, the failure is not necessarily intercrystalline but may be transcrystalline. In some observed cases a single specimen has shown both types of fractures within a negligible distance of each other.

In fatigue failures a metal after a number of repeated stress cycles of sufficiently high magnitude eventually suffers a localized damage and an incipient crack develops at a selected location. In such failures the path of the fracture is always transcrystalline. The inception of the fatigue crack is determined by the presence of minute discontinuities either within the metal or on the surface of the metal. The lowered fatigue value of metals when subjected to the additional element of corrosion is due to the production of sharp corrosion notches in the metal surface which serve as particularly damaging stress raisers. The path of fracture is again transcrystalline. All metals suffer lowered corrosion fatigue value as compared with corresponding fatigue values in air.

In the relatively limited number of cases of stress corrosion or "season cracking", a steady or static stress as distinct from repeated stressing is involved and with one exception the paths of such failures are intercrystalline. While such failures are localized they generally involve more than one fracture, the fractures developing at any location throughout the stressed area. Note, for example, the many centers from which cracks originated in the stressed columbium-bearing 18-8 sample of Fig. 16. In spite of these differences between stress corrosion or "season cracking" and corrosion fatigue failure the mechanism of the two phenomena is probably similar. The corrosive media produce sharp notches in the stressed metal surface which act as nuclei for the rapid propagation of the

cracks. Where the grain boundary material is more reactive with the corroding media than the matrix, the sharp corrosion notches are first formed in the grain boundaries at the surface. The grain boundary material is usually less resistant to corrosion than the grain itself because of segregation of impurities during freezing or during recrystallization, because of precipitation effects, or because of the existence of more disorganized and higher strained regions resulting from the opposing attractions of separate grains during crystallization or recrystallization. When, therefore, the material is stressed within a certain range and subjected to corrosive influences the fracture develops preferentially along the grain boundaries and the path of failure is intercrystalline. Since a pure stress failure is normally transcrystalline it would seem that in intergranular stress corrosion failure the influence of the corrosive medium must be the dominant factor, whereas in the transcrystalline corrosion fatigue fractures the corrosion element must be of only secondary importance.

The exception to the general intergranular type of stress corrosion failures, namely, the transcrystalline cracking of the austenitic chromium-nickel steels, may be explained by the increase in the stress magnitude along the slip planes of the grains due to the tendency of the austenite in these regions to transform under the influence of stress. The austenitic chromium-nickel steels are structurally unstable even when slowly cooled from high temperatures. The fact that severe cold working will produce an almost complete allotropic transformation of the austenite to ferrite is sufficient proof that this is so. The stressing of such an unstable alloy produces at the slip planes a tendency to transformation of ferrite with accompanying volume change, producing localized highly stressed regions. Stress corrosion cracks in austenitic nickel-chromium alloys, not susceptible to intercrystalline corrosion, are therefore transcrystalline by reason of the greater importance of the stress factor. For identical reasons transcrystalline cracks may also occur in material susceptible to intercrystalline corrosion.

## Summary

Service failures and experiments indicate that the austenitic chromium-nickel steels are susceptible to stress corrosion cracking and that such stress corrosion cracking may occur in such materials independent of its condition with respect to susceptibility to inter-

granular corrosion. The tendency to failure is, however, more pronounced when the material is in such condition that intercrystalline corrosion can develop.

The path of stress corrosion cracks in these alloy steels may be intercrystalline or transcrystalline, dependent upon the corrosive medium, the stress magnitude and the degree of susceptibility of the material to intergranular corrosion. The cases of transcrystalline cracking of austenitic chromium-nickel steels which have been noted are exceptional since other known cases of stress corrosion or "season cracking" of other metals are invariably intercrystalline.

Of many corrosive agents investigated, a mixture of ethyl chloride and water was found to be very effective in producing stress corrosion failure. Ferric chloride and mercuric chloride solutions were also found to have a very damaging effect. In both service failures and in experimental tests even mildly corrosive agents, which would leave plain carbon steel unaffected, were capable of producing stress corrosion cracking.

Tendency to stress corrosion cracking increases with increasing stress. Highly stressed conditions may be readily introduced by certain fabricating operations such as cold bending, welding or rapid cooling from elevated temperatures, and also by the use of composite structures, such as a welded attachment between austenitic and ferritic steels.

Elimination of the possibility of stress corrosion cracking can only be secured by removal of fabricating stresses by annealing treatments utilizing slow rate of cooling. This requires the use of stabilized alloys containing columbium or titanium to avoid the possibility of intergranular corrosion. Design of structure and thermal operation of structure in service should be such as to eliminate the introduction of high service stresses. If high service stresses should be introduced, there is no assurance that annealed structures of stabilized alloys will not fail under certain types of corrosive conditions, the field of which has not as yet been completely explored.

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## DISCUSSION

**Written Discussion:** By H. M. Wilten, metallurgist, The Texas Company, Port Arthur, Texas.

The authors ably presented a problem of stress corrosion cracking of the austenitic chromium-nickel stainless steel. The intergranular service failures of

these steels have been extensively studied and described. The transcrystalline type of stress corrosion cracking described herein, however, being not so well known, is considered "unusual" by the authors.

This type of cracking presents an important service problem in the use of these steels in the oil refining industry. The complete transcrystalline disintegration of exchanger tubes of 18-8 analysis has already been described

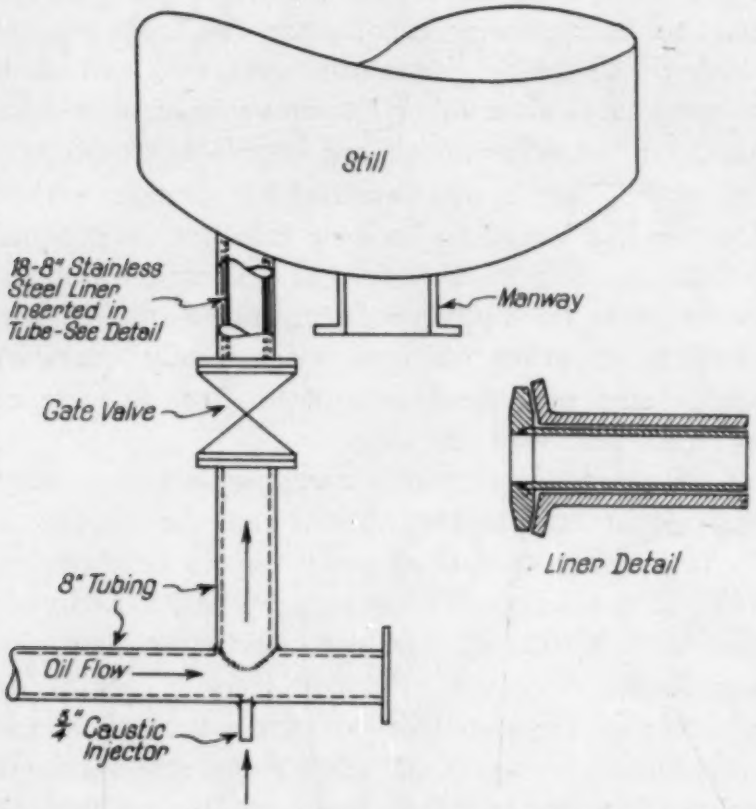


Fig. 1—18-8 Stainless Steel Liner, Its Detail and Mode of Installation.

(see Ref. 1, the text). The active corrodent in that case was probably some derivative of sulphuric acid with which the naphtha was treated and possibly not completely neutralized.

It may be of interest to present another instance of the same type of disintegration. This occurred in a liner. The detail of the liner construction and installation is shown in Fig. 1. The liner had been installed to protect the still nipple from corrosion by lubricating oil distillate at 460 to 500 degrees Fahr. at the location of injection of caustic. The liner was manufactured of ordinary 18-8 sheet ⅛ inch thick by cold bending in the press and by welding. The cylinder was also welded to the gasket. The liner was not heat treated in any manner. The analysis of the principal constituents of the steel was as follows: carbon 0.10 per cent, chromium 18.37 per cent, nickel 10.2 per cent. This liner was disintegrated after approximately two weeks service. Fig. 2 illustrates the cracking of the edge next to the weld. Fig. 3 illustrates the other end of the liner.

It is to be noted that the liner was not under any operating stress, only under internal stress resulting from bending. The active corrodent in this case was caustic. The oil itself is not thought to be responsible, as no instance of cracking was noted on many liners in service on cracking units at temperatures ranging from 100 to 1000 degrees Fahr.

Cracks originated mainly from the outside surface. Fig. 4 illustrates the etched portions of the liner. Only very few cracks started from the inside

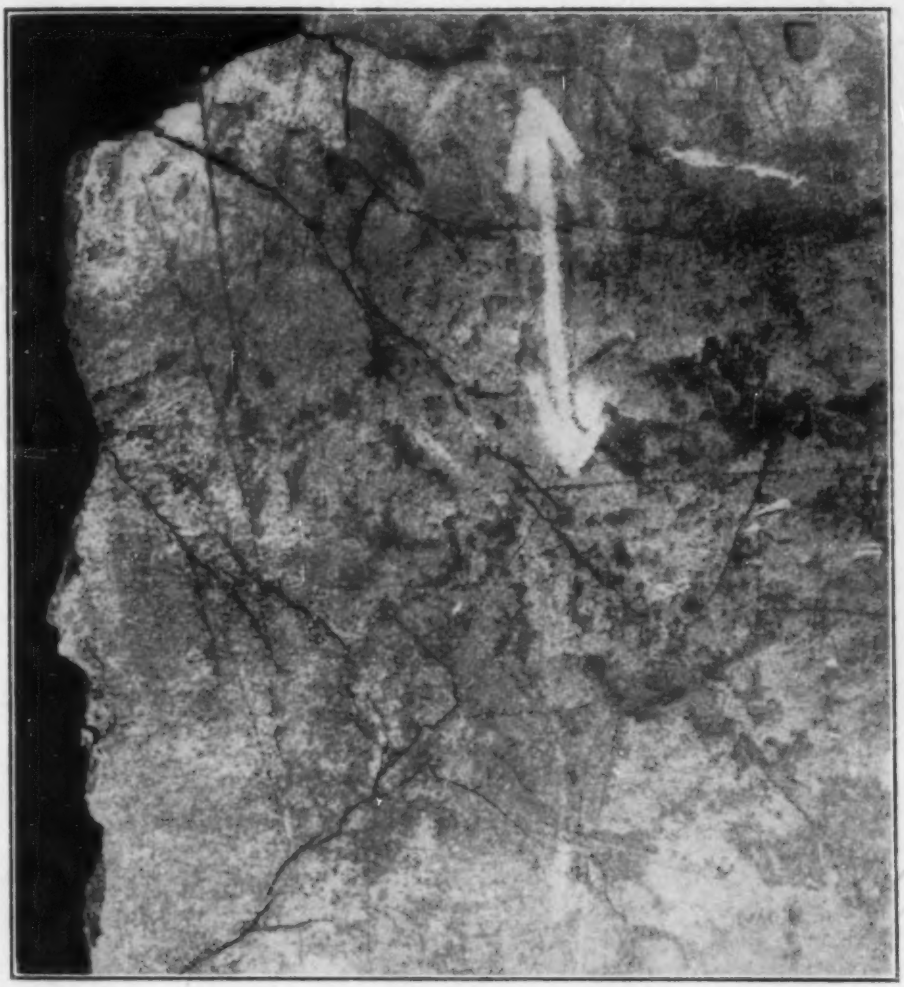

Fig. 2—Cracking of Liner at the Junction of the Weld.

surface of the liner. In all cases the cracking was transcrystalline, a typical crack being illustrated in Fig. 5. Further, it is interesting to note that transcrystalline cracking occurred, even though the metal in the liner was susceptible to intercrystalline corrosion, as a section of the liner cracked on bending after being boiled in the Strauss solution (10 per cent sulphuric acid and 10 per cent copper sulphate.)

This service failure was solved by manufacturing the liner out of Inconel, thus 18-8 stainless steel, in spite of its corrosion-resisting properties, might have been inadvertently blacklisted again. The authors' successful experi-

ments of eliminating failures by means of reduction of internal stresses in the metal lead us to believe that proper heat treatment with this point in view will solve this particular problem. An experiment to check this belief has already been started.

The authors are to be congratulated for this contribution to the science of metallurgy. The great importance of the stress factor which they so ably brought out will possibly explain many failures which were previously dis-

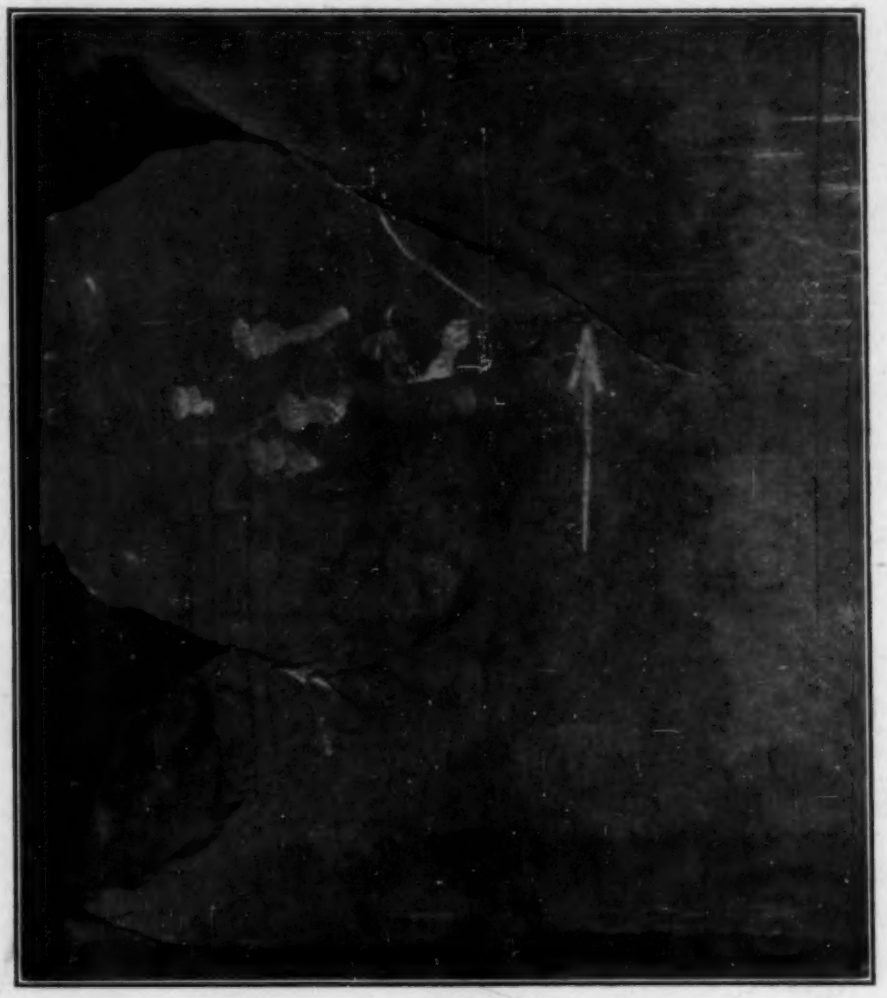

Fig. 3—Cracking of Liner on the Opposite Free End.

missed by interpreting these as "corrosion fatigue" failures, and thereby implying that service stresses were responsible for these failures. For example, many failures of austenitic 18-8 steel exchanger tubes were explained in that manner.

The author of this discussion will consider it successful if the authors of this paper, and possibly some of the readers, will cease thinking of transcrystalline cracking as being exceptional. Why not consider these failures as logical cases, wherein the importance of internal stress has not been adequately taken into account?

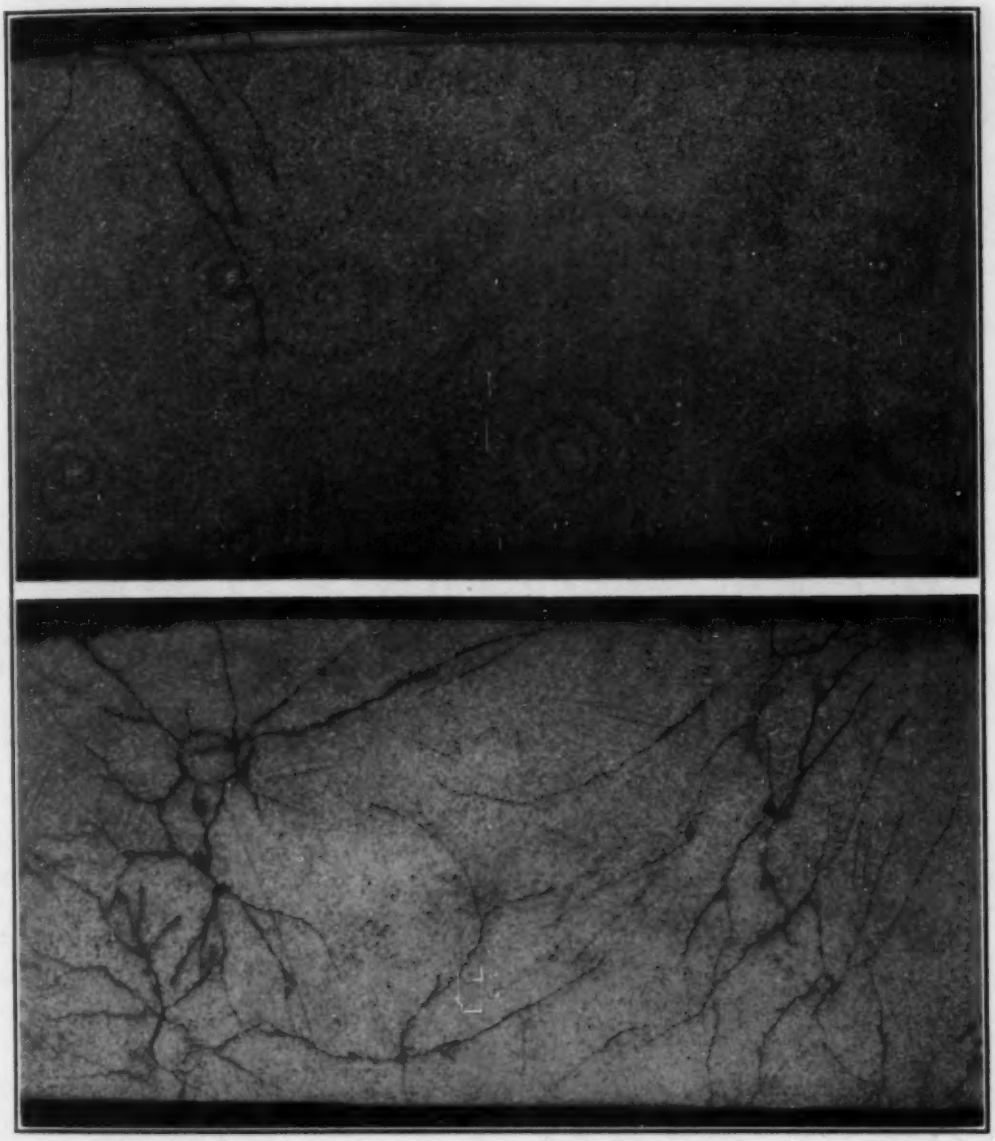

Fig. 4—Etched Sections of the Liner Illustrating Cracking. The Upper Section is that of the Outside Surfaces. The Lower Section is that of the Inside Surfaces.

**Written Discussion:** By F. R. Steckel, metallurgist, Columbia Steel Co., San Francisco.

This paper is of special significance as it indicates that static stress is a much more important factor in the corrosion of austenitic stainless steels than has generally been suspected. The static stress factor has often—perhaps we should say usually—been totally ignored in corrosion investigations. Conflicting data developed in investigations made in the past have often defied explanation, but perhaps in light of this new knowledge, regarding the influence of static stress, facts which appear to be in conflict may now be reconciled.

It is trusted that before long there will be a full investigation to determine the effects of all kinds of corrosive agents on stress corrosion cracking. The authors' work shows definitely that ethyl chloride, mercuric chloride and fer-

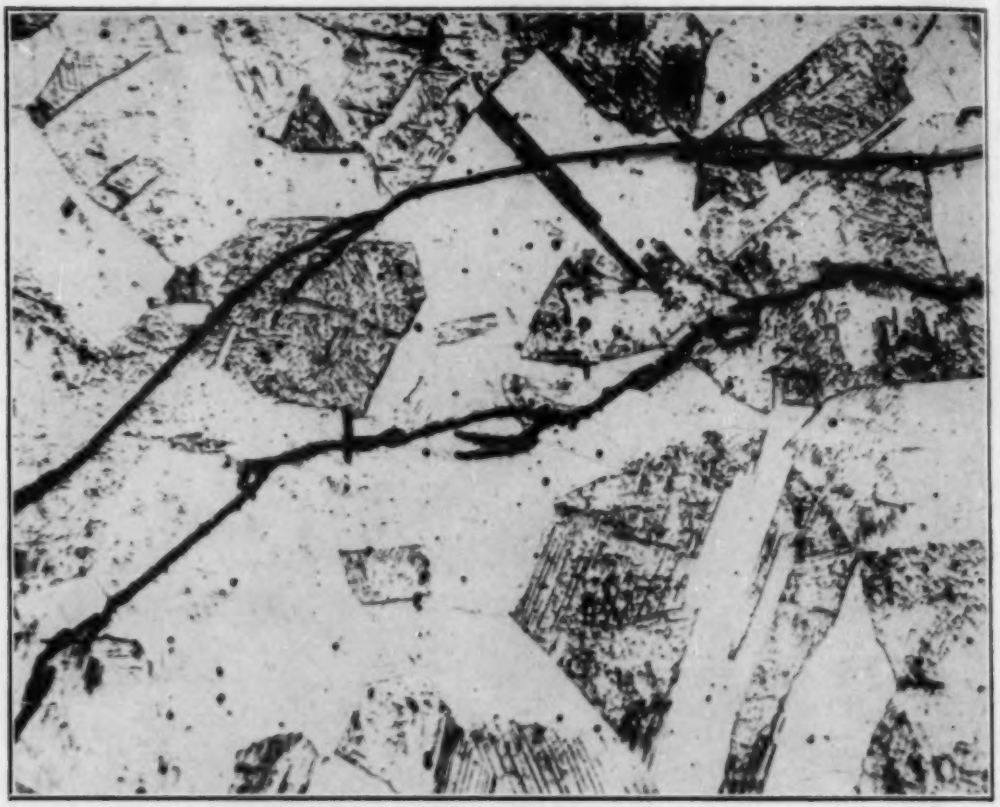

Fig. 5—A Typical Transcrystalline Crack. Etch Glyceregia. × 250.

ric chloride solutions are particularly active in causing this type of failure, and while the statement that the corrosive media capable of causing stress corrosion cracking in austenitic chromium-nickel steels are limited in number is reassuring, it will lead to serious doubts in the minds of many people as to possible stress corrosion effects when dealing with some particular corrosive agent or condition.

Practically all fabricated assemblies contain locked-up stresses in some degree. Stress relieving is probably employed in fewer instances with austenitic stainless steels than with other ferrous alloys. This is because of the fact that these austenitic stainless steels are very ductile non-air-hardening materials, and there is comparatively little danger of cracking at the welds due to forces of contraction. It now appears that these internal stresses are important, not only from the standpoint of strength, but as they affect corrosion resistance.

The conclusions reached in the authors' paper may well lead the way to more stress relieving than has been done heretofore. This emphasizes the importance of obtaining a great deal of data regarding the effects of many different kinds of corrosive media.

The thought occurs to us that perhaps many corrosion failures attributed in the past to carbide precipitation may be due, in part at least, to internal stresses. The so-called heat affected zone which runs adjacent to and parallel to a welded seam is also the zone where internal stresses would reach their highest magnitude.

**Written Discussion:** By Merrill A. Scheil, research metallurgist, A. O. Smith Corporation, Milwaukee.

I have read with great interest the authors' laboratory experiments which duplicated the service failures of thermowells made from 18-8 molybdenum and 18-8 stainless steel. It is apparent from their photograph, Fig. 8A, that the weld metal used to fabricate the 18-8 molybdenum thermowells can fail by cracking after a quench from 2000 degrees Fahr. It would be of interest to learn the chemical analysis of the weld metal used to fabricate thermowell A, which was not heat treated prior to testing, and that used for the thermowell of Fig. 7B, which was heat treated, as both the welds failed. Photographs Fig. 7A and 7B do not duplicate the sketch of Fig. 3 showing the carbon steel capillary tube welded to the plug of 18-8 molybdenum. Was this detail omitted in the laboratory tests? It is also not clear why the bore of the plug and the outside surface of same do not show corrosion cracks on photograph 7B. It would be of interest if the authors would explain this. Was the chemical composition of the plug material similar to that of the forged and machined tube?

Table I gives two typical analyses of thermowells that failed in service. It is not clear whether the experimental thermowells which were fabricated and tested were all of the same analysis as the failed materials. In this connection it would be interesting to learn whether the authors used a material corresponding to a type 317 analysis for any of their experiments.

It would be of particular interest if the authors would comment more fully on the number of analyses or heats which they have tested in their experiments to determine the quality or characteristics of materials that are susceptible to stress corrosion cracking.

While this paper is interesting, it may be misinterpreted and is therefore liable to create a negative reaction to austenitic steels in any form, unless it is realized that the tests, as carried out in the experiments, relate to a selected media which may not be duplicated in many types of service or even in the service described by the authors when moisture is eliminated from the ethyl chloride. Our company has been successfully lining a great variety of vessels with austenitic materials for a wide diversity of service over a period of the past ten years. We know from actual experience, as well as from our laboratory work, that these materials when properly selected and properly handled can be used under highly stressed conditions without reducing their effectiveness in a great many cases.

**Written Discussion:** By W. B. Brooks, metallurgist, Stainless Steel Division, Carnegie-Illinois Steel Corp., Pittsburgh.

This is a very distinct contribution to the literature on the corrosion of stainless steels. The data clearly establish a phenomenon which has caused some confusion and one which at times has not been clearly recognized and understood. The authors have very ably demonstrated that a considerable static stress, when accompanied by the presence of certain chlorides or conditions favoring intergranular corrosion, will cause cracking. It is the purpose of this discussion to re-examine the implications which may be drawn from the data.

The suggestion is made that these failures may be closely allied to the season cracking of brass and other face-centered cubic metals and alloys. Sea-

son cracking, as it is commonly understood, is a purely intercrystalline phenomenon whereas here we are dealing with transcrystalline as well as intercrystalline cracks. It is believed that the data may be clearly explained and that no direct relationship between this type of cracking and season cracking may be established.

Some alarm is felt over the fact that these failures were caused by ethyl chloride, which is not a notoriously corrosive material. It is stated that carbon steel under the conditions of tests remain totally unaffected. The point is that the corrosive was not ethyl chloride, per se, but a mixture of ethyl chloride and water under conditions favoring the formation of hydrochloric acid. The susceptibility of the stainless steel to localized attack, followed by various chlorides, is well known and failure takes place by pitting. It is unfortunate that no direct weight loss comparisons between the plain carbon steel and the 18-8 were made since it is quite probable that the plain steel would have shown a somewhat greater loss. In any case the weight loss of the stainless steel suffering this type of attack is extremely small, probably a small fraction of that incurred by carbon steel under the same conditions.

No corrosive will of itself cause a transcrystalline crack. Only stress will cause such a crack. Now the mechanical properties of these alloys are well known. Depending on the nature of their application, stresses in the order of 60,000 to 100,000 pounds per square inch are required to cause a rupture. Such overall stresses do not exist in the structures under consideration or they would be severely deformed, if not actually ruptured, very quickly. The stresses involved here cannot be more than slightly in excess of the yield point, yet we are confronted with transcrystalline cracks. How may these cracks be explained?

The mere combination of stress and corrosion is not sufficient. This is well demonstrated in the investigation reported by Hoyt and Scheil.[1] Their investigation was largely confined to studying a mixture of nitric and hydrofluoric acid, a mixture which causes general surface solution rather than sharply localized attack as in the case of chlorides. As a matter of fact, industry is full of conditions involving the co-existence of stress and corrosive conditions. Evidently, specific corrosive conditions must be present. Have we any clues as to the nature of these conditions? The answer is decidedly yes. They must, as the authors suggest, be of such a nature as to cause points of stress concentration—in other words, notches. The ability of the corrosive conditions outlined in this paper to do that may be easily recognized. In one case, we are dealing with chlorides which cause pits, and in the other, with intercrystalline corrosion which also provides notches in the initial grain boundary attack. Thus the simultaneous occurrence of inter and transcrystalline cracks may be readily explained. This notch effect may be appreciated by a glance at a stress distribution diagram shown in Fig. 6.

The formation and propagation of the crack are favored by the mechanics of the situation. Fig. 7A shows the development of a pit in the absence of stress. Fig. 7B shows the development of a crack from a pit when stress is present.

In pits occurring in the absence of stress, the corrosion actually shows a

[1] S. L. Hoyt and M. A. Scheil, "Stress-Corrosion Cracking in Austenitic Stainless Steels," TRANSACTIONS, American Society for Metals, Vol. 27, 1939, p. 191.

tendency to wander and we have the familiar undercut pinhole. On the other hand, with stress present, the point of maximum stress concentration (the bottom of the pit) provides a point of minimum corrosion resistance; the cor-

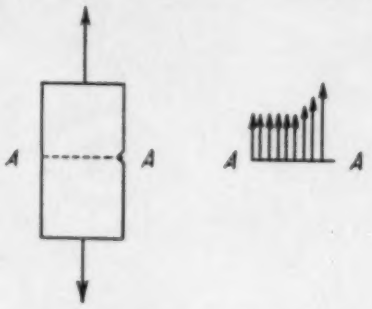


Fig. 6—Stress Distribution Diagram Showing the Effect of a Notch.

rosion is sharply localized at this point and progresses rapidly inward normal to the direction of the stress, giving an increasingly sharper notch which soon becomes a crack.

This same analysis may be applied to the laboratory and service failures associated with susceptibility to intergranular corrosion. Fig. 8.

It is to be remembered, in connection with the propagation of transcrystalline cracks in material which has been subject to carbide precipitation, that

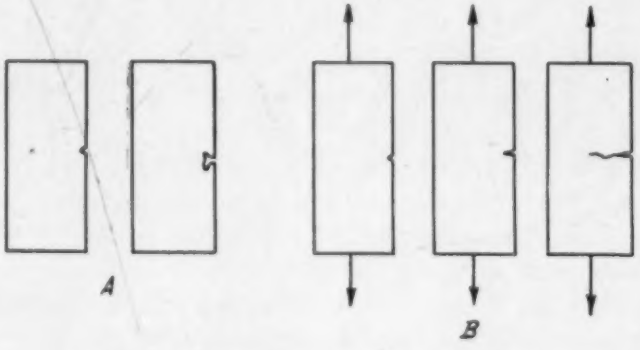


Fig. 7A—Illustrating the Development of a Pit in the Absence of Stress.
Fig. 7B—Illustrating the Development of a Crack from a Pit when Stress is Present.

this precipitation may also take place on slip planes and on the boundary of twins. Intercrystalline corrosion, its cause and prevention are now well understood.[2]

To summarize, the conditions which cause this cracking must provide points of stress concentration either by pitting or by intercrystalline attack. Fortunately, these conditions are widely recognized and may be avoided. Intercrystalline corrosion should no longer be a matter of concern. It is prevented by heat treatment or by the addition of titanium or columbium to 18-8.

[2]E. C. Bain, R. H. Aborn and J. J. B. Rutherford, "The Nature and Prevention of Intergranular Corrosion in Austenitic Stainless Steels," TRANSACTIONS, American Society for Steel Treating, Vol. 21, 1933, p. 481.

The conditions which cause pitting are also well known and are carefully avoided by industry. It would not seem, therefore, that we may anticipate failures of this type occurring in unsuspected places, so the industrial implications of this phenomenon do not loom large. Failures are rare and will become

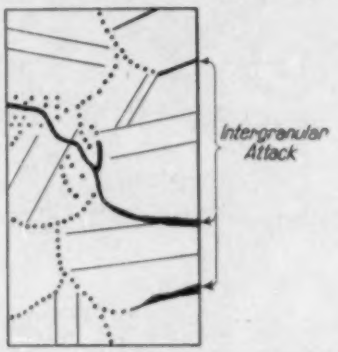


Fig. 8—Schematic Drawing Showing Mechanism of Intergranular Attack.

more so as understanding spreads. We are, indeed, indebted to Messrs. Hodge and Miller for their work, which so ably supplements that of Hoyt and Scheil, and which adds a valuable contribution to our knowledge of corrosion.

**Written Discussion:** By the Research Staff of Allegheny Ludlum Steel Corp., Brackenridge, Pa.

From time to time we become aware of the failures of metallic materials under apparently inexplicable conditions. Invariably these conditions involve the presence of stresses (often elastic) and mild corrosive media. It may happen either at room or elevated temperature. Such failures should have received a great deal more of attention than they have had. Fundamental factors, back of the phenomenon, are not clearly understood; in fact, even the nomenclature on the subject is not clearly established. It may be, as we suspect, due to experimental difficulties: any problem involving internal stresses is a complex one. Our lack of concentrated attention may be explained on the ground that failures analogous to those reported in this paper are far from being a common occurrence. Further, it is known that in many instances the danger has been completely removed by the change in the design or by some other practical means. Thus, a fundamental and comprehensive study of the phenomena has not been imperative.

Concrete data which are to be found in the work of Messrs. Hodge and Miller contribute to our collection of actual facts—for this reason alone it should be welcomed. In addition it contains interesting experimental findings which may be found invaluable in the future comprehensive studies.

In the course of the paper the authors state the necessary conditions under which failure may occur. These conditions are (1) the presence of stresses of which the authors say that "it is the primary requisite of the occurrence" (i.e., failure) and (2) the presence of corrosive agent on which—in this particular case ethyl chloride—the authors lay the blame—calling it the "active agent". There is no doubt that such stipulations are correct—the interesting evidence is there. The interpretation of the results and especially the interpretation of laboratory results may be attempted in somewhat different light.

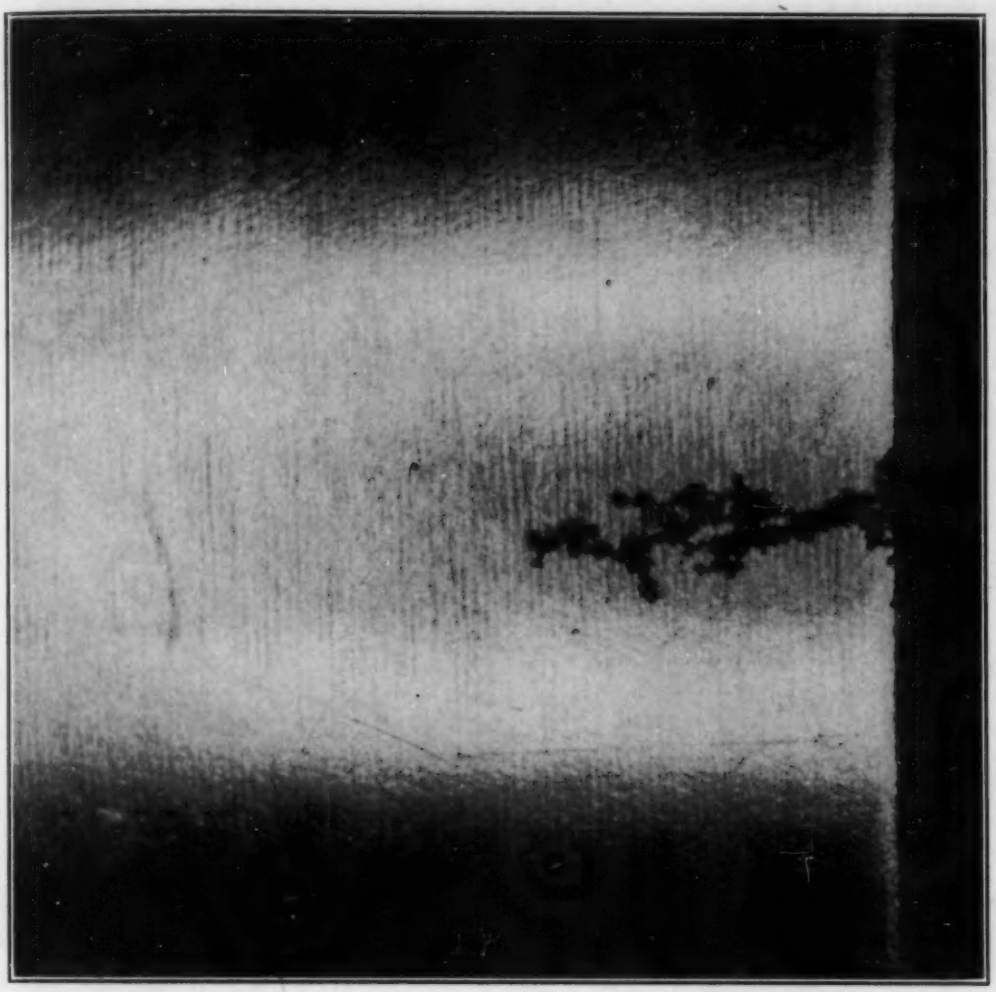

Fig. 9—Pinhole Corrosion in Chromium-nickel Austenitic Alloy. Sample Under Stress. × 8.

Stress corrosion is a somewhat indefinite term: it may and often does include corrosion fatigue. This is borne out by the fact that any stress range, no matter how low, can cause failure under corrosive conditions, if given indefinite time. It does not matter, either, if the stress fluctuates within narrow limits. Neither is so important as the frequently observed fact that the damage from *simultaneous action* of stress *and* corrosion greatly exceeds that when these two agents are operating separately. Professor Gough expresses it rather well and explains it as follows: "First of all," he says, "the two effects are not merely additive. In the usual type of corrosion test, the specimen may be subjected to stress but the conditions are essentially *static;* it is common experience to find that the presence of the resulting corrosion products tends to retard the attack or to cause it to cease completely. Directly cyclic *stresses* are superimposed, the induced cyclic *strains* tend to rupture, or render more permeable the protective film and to dislodge or remove the other products of corrosion. Hence, the cyclic stresses may be said to accelerate corrosion in an indirect manner."

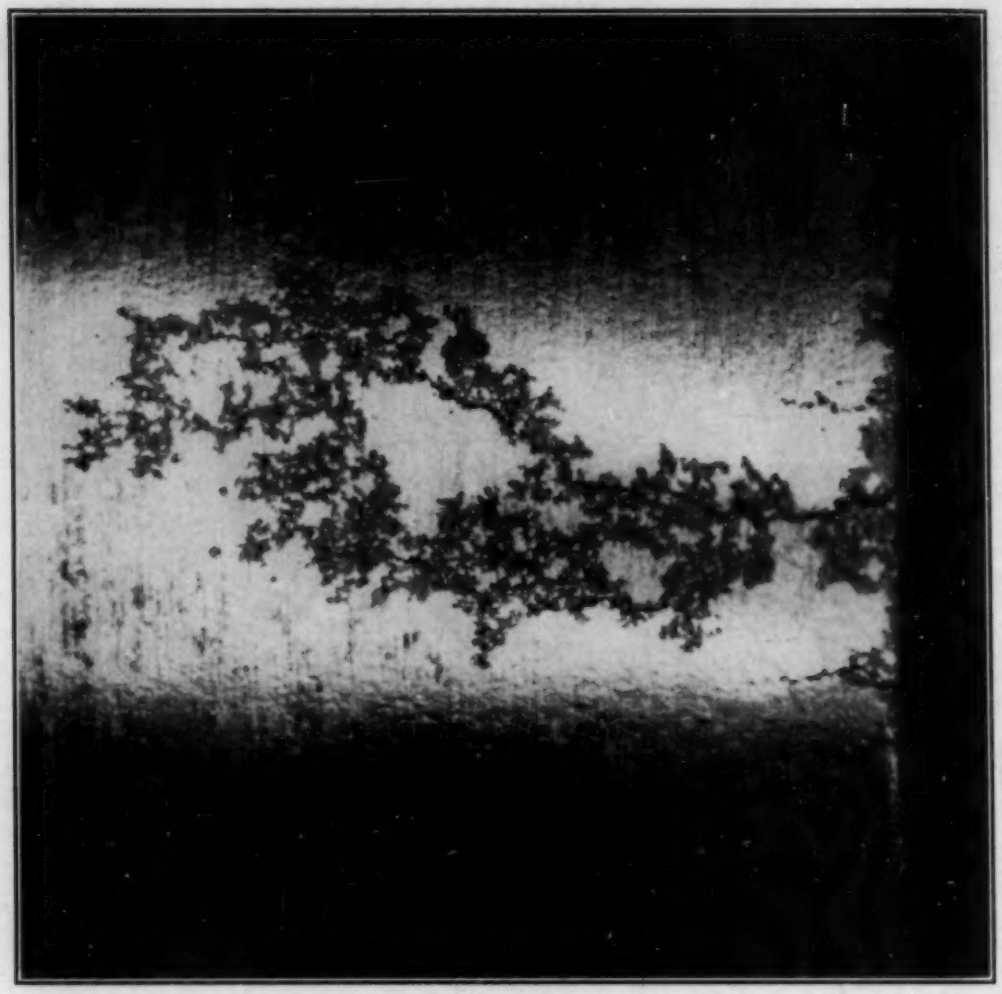

Fig. 10—Stress Corrosion in Stressed Chromium-nickel Austenitic Alloy. × 8.

In the experimental work of Messrs. Hodge and Miller on thermowells with ethyl chloride the possibility of corrosion fatigue is not excluded. They state (on page 44) that the pressure within thermowells was changing from 400 to about 6000 pounds per square inch. Whether the fluctuations in stress were slow or rapid (as we usually find in fatigue) does not appear to be vastly important. We believe that in either case the explanation of Gough would apply. It is worthy of mention here that when in a certain installation a case of corrosion fatigue was discovered the same was effectively remedied by allowing free expansion and contraction of the metal.

In studying the phenomenon of corrosion and stress a constant watch over the sample under observation yields interesting facts. For example, a thin sample of chromium-nickel alloy stressed by bending and subjected to corrosive action of ferric chloride showed first the well known "pinhole" attack, the pinholes being bunched in a cluster (Fig. 9). No cracking was to be seen. After a more prolonged exposure a definite cracking (in a similar sample and place) had occurred—the sample no longer maintained its original shape, but, because of its thinness, became distorted (Fig. 10). This is exactly what we should expect because pitting of the surface by corrosion gives rise to mechanical stress concentration effect.

The experience of the authors with failure in superheater tubes has not been evaluated in the light of very recent work on "Fractures of Steels at Elevated Temperatures After Prolonged Loading", recently reported by Thieleman and Parker and published in *Metals Technology*. These authors observed brittle fractures at elevated temperatures and after much comprehensive study showed that most of the investigated steels may be made to fail through brittle fracture, providing the conditions of temperature and stresses are properly selected. The role of possible corrosion (in this case oxidation) does not appear to be clearly determined: the authors found that intergranular, i.e., brittle fractures may or may not be accompanied by intergranular oxidation.

As we believe we already said the report of Messrs. Hodge and Miller adds very considerably to our factual information. To quote the authors, "The corrosive media capable of producing stress corrosion cracking in austenitic chromium-nickel steel are limited in number." Even though it is so the effect of stress—even in the elastic range—on intergranular and overall corrosion of iron, brass, aluminum, magnesium, duralumin, etc., deserves the attention of scientific workers. It is sincerely hoped that these authors will continue their work on this important subject.

**Written Discussion:** By A. E. White, director, Department of Engineering Research, University of Michigan, Ann Arbor, Mich.

We are all deeply grateful for the extremely helpful paper by Messrs. Hodge and Miller on "Stress Corrosion Cracking of the Austenitic Chromium-Nickel Steels." The paper to my knowledge covers an immense amount of research undertaken to explain what were, at the time, some baffling failures of 18-8 material.

Many have felt that 18-8 stainless steel, with certain elementary precautions, was relatively fool proof. This paper explodes this viewpoint. It issues a direct warning to the fabricators of this material that the alloy must be so assembled as to be completely free from fabricating stresses; which means annealing treatments utilizing slow rates of cooling. As such treatments possibly lead to intergranular corrosion the alloy should be treated with columbium or titanium to reduce to the minimum the tendency to failure from this cause. Also care must be taken that the service stresses are not unduly high.

**Written Discussion:** By Clair Upthegrove, professor, department of chemical and metallurgical engineering, University of Michigan, Ann Arbor, Mich.

Messrs. Hodge and Miller have presented an interesting and informative paper on the question of failures in austenitic chromium-nickel steels. Emphasis has been placed upon the presence of internal stress and contact with a corroding medium as necessary factors for the development of inter or transcrystalline cracking of the stainless steel and ultimate failure of the part involved. It appears doubtful though that the occurrence or nonoccurrence of the cracking in the stainless steels and the occurrence of the cracks as intercrystalline or as transcrystalline is determined by only the above two factors. In fact the authors suggest a third factor in their summary when they speak of the degree of susceptibility of the material to intergranular corrosion. May there not be also a degree of susceptibility of the material to transcrystalline cracking as well as to intercrystalline cracking? In the studies made using

the various bottles which were capped and tested with ethyl chloride-water mixture, it may be assumed that the corroding medium is constant and that only variations in internal stress are a factor in causing cracking. A comparison of Bottles A and B will show that no cracks were found in the weld heat affected portions of the bottom cap of Bottle A but that cracks were present in the weld heat affected portion of the bottom cap of Bottle B. In this case the same weld procedure and the same corrosion medium produce cracks in one 18-8 cap but not in the other. In the same way, it may be assumed that while the complete procedure varied for Bottles C, D and E, stresses set up in the arc welding of the lower 18-8 cap to the 18-8 tube section should be the same in all cases for that particular weld section. Bottles C and E were stress relief annealed at 1350 and 1650 degrees Fahr., respectively, and cooled slowly to room temperature. Bottle D was stress relief annealed at 1350 degrees Fahr., but was furnace cooled only to 600 degrees Fahr., after which it was air cooled to room temperature. May it not be reasonably assumed that while the weld section, carbon steel to 18-8 tube, unlike materials, would be affected by the differences in their coefficients of expansion, the lower weld section of 18-8 tube to 18-8 cap, like materials, would not be affected by the differences existing at the middle weld section? Might it not have been expected that maximum stresses would have been produced by the rapid cooling at or near the middle weld section of Bottle D and that cracking should have occurred to the greatest extent at that point rather than as indicated in the sketch at the lower weld section? While the presence of stresses and corroding media are necessary for cracking of the stainless steel, other factors may apparently contribute to the origin or development of the cracks.

**Written Discussion:** By R. H. Aborn, research laboratory, United States Steel Corp., Kearny, N. J.

The authors are to be commended for this extensive work which illuminates what may well prove to be the one remaining peculiarity of austenitic stainless steels.

It seems evident that the path of stress-corrosion cracking in austenitic stainless steel tends to be *intergranular* only when the material is susceptible to intergranular corrosion. The specimens described by Hoyt and Scheil (TRANSACTIONS of American Society for Metals, Vol. 27, 1939, p. 191) were in this condition and behaved accordingly.

It appears that *transgranular* corrosion cracking may occur when two factors are present: high internal stress and immersion in an aqueous solution containing chloride ions. Thus far, this is the only combination which has been found to produce the phenomenon, and apparently only within certain limits of composition and possibly of stress is the combination effective.

Since it is well recognized that the chloride ion under favorable conditions causes pitting, it seems likely that a pit possibly of microscopic dimensions is the nucleus of a transgranular crack which develops under an appropriate stress. Figs. 16 and 11 are strongly suggestive of such a process.

The majority of applications of austenitic stainless steels do not involve high internal stress combined with immersion in an aqueous chloride solution, and of those which do, experience indicates that the necessary critical condi-

tions are usually not present. In doubtful instances as well as in those of the type described by the authors, the prospective user will do well to use titanium or columbium-bearing alloys and after fabrication is completed, apply a stabilizing stress-relief anneal followed by slow cooling.

### Authors' Reply

We were very much interested in the information concerning two service failures involving transcrystalline stress corrosion cracking as described by Mr. Wilten. In both of these cases chlorides were evidently absent, thus indicatting that under special circumstances other substances will produce similar failures. It is notable that these transcrystalline cracks developed in spite of susceptibility to intercrystalline attack resulting from carbide precipitation. This corroborates our experimental results which showed that such failures may be either intercrystalline or transcrystalline dependent upon the relative importance of several factors.

As Professor Upthegrove has pointed out, there exist other factors of importance in influencing such stress corrosion failures, which we have not as yet been able to consider. One such possible variable is grain size and grain shape and another is variation in chemical analyses although it is believed that the latter is of little importance within fairly wide limits.

We concur with Mr. Brooks' conception of the mechanism of reaction necessary for the initiation of cracking. In the "General Discussion" the following statement appears: "The corrosive media produce sharp notches in the stressed metal surface which act as nuclei for the rapid propagation of the cracks". Whatever the mechanism of crack inception may be, the evidence indicates that the combination of static stress and corrosion *is* sufficient to produce localized cracking in the austenitic steels tested, under circumstances which produced no localized cracking in carbon steels or in stress-free austenitic steels.

In reply to the questions raised by Mr. Scheil we wish to comment as follows: the weld metal of the experimental thermowells was of approximately the same analysis as the bar stock used for the upper and lower portions of the thermowells. The experimental thermowells described in the paper were, as stated, made from the 18 per cent chromium, 8 per cent nickel, 3 per cent molybdenum analysis, a typical example being as indicated in Table I. The carbon steel capillary tube was not used in the experimental thermowells. The upper part of the thermowell which nests in the lower section is loosely fitted except where joined by welding. Thermal stresses of any magnitude could not, therefore, be induced except at the welded portion. In the as-welded specimens transcrystalline cracking did occur in this region. Other experiments not reported in the paper using 18-8 not containing molybdenum gave identical test results when treated as were Thermowells A, B and C. Approximately ten heats of steel were involved in the service thermowell failures and sixteen heats of steel were involved in the superheater tube failures. We have observed and investigated similar failures in many other service heats which covered a rather wide chemical analysis range.

We have welcomed the interesting discussion by the research staff of the Allegheny Ludlum Steel Corporation. We agree that the failures described should be considered as corrosion fatigue failures but in a very special sense. The meaning of the expression "corrosion fatigue" must be enlarged to include static as well as dynamic stresses.

In closing we wish to express our appreciation of the interesting comments submitted by Messrs. Aborn, White and Steckel.

# HARDNESS CHARACTERISTICS OF SOME MEDIUM CARBON S.A.E. STEELS

By P. Klain and C. H. Lorig

*Abstract*

*The hardening characteristics of sixteen S.A.E. plain carbon and alloyed, medium carbon steels, as revealed by a definite testing procedure, were examined on quenched, and on quenched and tempered specimens ½, 1, 2, and 3 inches in diameter. The specimens for quenching were 7 inches long. Eight of the steels were quenched in water, the remainder were quenched in oil having a viscosity of 100 Saybolt seconds at 100 degrees Fahr. Quenching temperatures were 50 degrees Fahr. or more above the $Ac_3$ point for each steel, and as a rule, were those recommended in the S.A.E. Handbook for the steels.*

*The rate of flow of both coolants through the quenching fixture was 105 feet per minute, a rate sufficient to obtain maximum surface hardness on 3-inch diameter specimens of the most shallow hardening of the steels tested.*

*A duplicate set of specimens for all four sizes of each steel was quenched. Only the center 3 inches of each specimen, cut into three 1-inch slices, was utilized in the hardenability studies. The middle 1-inch slice was explored for transverse hardness after removing from the surface the overheated layer formed in cutting off. All three slices were used in determining the hardness of the quenched steels after tempering.*

*The results of the study are summarized graphically in a group of figures showing conventional hardness penetration curves and tempering curves, the latter plotted to show the change in hardness, for different tempering temperatures, at the surface, at half radius and at the center of specimens of the different diameters, and to show the change in hardness across the section for each tempering temperature.*

THE depth hardening properties of steels have aroused wide interest among consumers and producers of steel. Such interest was evidenced through the excellent group of papers and discussions presented in the Symposium on Hardenability at the 1938 American Society for Metals Convention. Even in this comprehensive dis-

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. The authors, P. Klain and C. H. Lorig, are metallurgists, Battelle Memorial Institute, Columbus, Ohio. Manuscript received June 26, 1939.

cussion only meager hardenability data on standard S.A.E. steels were presented. A need for more data was pointed out by several discussers, since available hardenability information on these widely used steels is astonishingly scanty.

In view of this need the American Society for Metals Handbook Committee's Sub-Committee on Alloying Elements planned an experimental study. The co-operation of Battelle Memorial Institute, American Steel and Wire Company, Bethlehem Steel Corporation, Carnegie-Illinois Steel Corporation, Climax Molybdenum Company, Republic Steel Corporation, Timken Roller Bearing Company and Vanadium-Alloys Steel Company made it possible to undertake the work.

The first part of the program, now completed and reported upon in this paper, deals with the depth hardening properties of some medium carbon S.A.E. steels, as shown by hardness determinations on quenched and on quenched and tempered specimens.

The purpose of this investigation was to examine the hardening characteristics of the steels as revealed by a definite testing procedure and to indicate the differences which prevail among the common S.A.E. steels. For that reason, great care was taken to standardize the method for conducting the hardenability studies. In order to test under conditions resembling normal commercial practice, it was decided that all steels be heated in the normal atmosphere of the furnace to allow scaling conditions similar to those existing in commercial heat treating. Furthermore, the quenching temperature of the 2 and 3-inch diameter bars was arbitrarily raised 50 degrees Fahr. higher than the temperature used on the ½ and 1-inch diameter bars for any one steel. An important variable is the velocity of impingement of the quenching medium upon the surface of the specimens. An attempt was made to find the usual commercial range, but the quenching operations in practice were so varied, ranging from spray quenching to slow agitation, that the rate of flow of the coolant against the surface of the specimen was arbitrarily chosen so as to obtain maximum surface hardness on 3-inch diameter bars of the most shallow hardening of the steels tested.

## Steels Used

A group of sixteen S.A.E. steels, furnished by four steel producers, was tested in ½, 1, 2, and 3-inch diameter sections. Eight

**Table I**
**Composition, Grain Size, Quenching Temperatures, and Methods of Processing of the S.A.E. Steels**

| S.A.E. Type | Per Cent C | Mn | P | S | Si | Cr | Ni | Mo | V | McQuaid-Ehn Grain Size | Quenching Temp. Degrees Fahr. | Processing Method |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| **Water Hardening Steels** | | | | | | | | | | | | |
| 1045 F | 0.435 | 0.72 | 0.018 | 0.032 | 0.20 | .... | .... | .... | .... | Fine | 1475-1525 | Forged |
| 1045 C | 0.44 | 0.75 | 0.019 | 0.025 | 0.25 | .... | .... | .... | .... | Coarse | 1475-1525 | Forged |
| 2330 | 0.30 | 0.69 | 0.016 | 0.020 | 0.24 | .... | 3.47 | .... | .... | 6 | 1475-1525 | Forged |
| T-1330 | 0.34 | 1.62 | 0.023 | 0.023 | 0.24 | .... | .... | .... | .... | 6 to 8 | 1550-1600-1625 | Rolled |
| 3130 | 0.31 | 0.62 | 0.015 | 0.022 | 0.21 | 0.62 | 1.23 | .... | .... | 6 | 1500-1550 | Forged |
| 4130 | 0.30 | 0.70 | 0.022 | 0.026 | 0.23 | 0.69 | .... | 0.19 | .... | 6 | 1575-1625 | Forged |
| 5130 | 0.30 | 0.74 | 0.018 | 0.023 | 0.20 | 0.87 | .... | .... | .... | 6 to 8 | 1575-1625 | Forged |
| 6130 | 0.33 | 0.70 | 0.036 | 0.036 | 0.23 | 1.10 | .... | .... | 0.22 | 6 to 8 | 1575-1625 | Rolled |
| **Oil Hardening Steels** | | | | | | | | | | | | |
| 2345 | 0.45 | 0.75 | 0.008 | 0.012 | 0.23 | .... | 3.61 | .... | .... | 6 | 1450-1500 | Forged |
| 3145 | 0.46 | 0.87 | 0.022 | 0.027 | 0.27 | 0.68 | 1.14 | .... | .... | 6 | 1450-1500 | Forged |
| 3240 | 0.39 | 0.49 | 0.020 | 0.020 | 0.25 | 1.15 | 1.69 | .... | .... | 6 to 8 | 1500-1550 | Rolled |
| 4145 | 0.45 | 0.80 | 0.020 | 0.027 | 0.16 | 0.94 | .... | 0.17 | .... | 6 to 8 | 1575-1625 | Forged |
| 4340 | 0.38 | 0.64 | 0.014 | 0.017 | 0.27 | 0.72 | 1.72 | 0.34 | .... | 7 to 8 | 1500-1550 | Rolled |
| 4645 | 0.46 | 0.65 | 0.014 | 0.017 | 0.21 | .... | 1.72 | 0.23 | .... | 5 to 8 | 1500-1550 | Rolled |
| 5145 | 0.455 | 0.84 | 0.018 | 0.021 | 0.19 | 1.00 | .... | .... | .... | 6 to 8 | 1575-1625 | Forged |
| 6145 | 0.49 | 0.71 | 0.022 | 0.022 | 0.21 | 0.95 | .... | .... | 0.21 | 6 to 8 | 1575-1625 | Rolled |

of these steels were for oil hardening and eight for water hardening tests. The type of steels used, together with their chemical analyses, grain size ratings, and quenching temperature are given in Table I. All four bars of each steel were forged or rolled (see Table I) from the same heat, $\frac{3}{16}$-inch oversize on diameter. They were then normalized ½ hour at 1700 degrees Fahr., tempered for satisfactory machinability, and turned to proper size.

## Testing Procedure—Quenched Bars

*Size of Specimens*—The specimens for quenching consisted of ½, 1, 2, and 3-inch rounds, 7 inches long. Only the 2 and 3-inch bars were drilled and tapped at one end for the insertion of $\frac{5}{8}$-inch bolts which made for more convenient handling of the specimens from the furnace to the quenching fixture. Two specimens of each bar size of each steel were quenched.

*Finish*—No attention was given to the surface finish of the specimens. Some had coarsely ground while others had rough machined surfaces.

*Selection of Quenching Temperature*—To establish the proper quenching temperature for each steel, it was intended to compare the fractures for a set of ½-inch diameter bars quenched at temperatures differing by 50 degrees in the range from 1450 to 1600 degrees Fahr. (785 to 870 degrees Cent.). The temperature giving

the finest fracture for a steel was to be selected as the proper quenching temperature for that steel. No clear-cut distinction between fractures could be observed, however, and this method was abandoned.

The temperatures finally selected were at a minimum of 50 degrees Fahr. above the $Ac_3$ point for each steel. In some cases it was considerably higher, especially when the steels contained difficult soluble carbides. As a rule the same temperatures as are recommended in the S.A.E. Handbook were employed. Quenching temperatures for each steel are included in Table I.

The arbitrary selection of higher temperatures for the 2 and 3-inch bars was made to offset the effect of mass on the hardening characteristics of steel when quenched in different sizes from the same temperature. This method is generally employed in commercial practice. It was appreciated that raising the temperature 50 degrees Fahr. might change the austenitic grain size, but in this case the temperature for the larger sizes was selected by consideration of hardenability and not from the standpoint of grain refinement of the steel.

*Quenching Media*—Water and oil were used as coolants. The temperature of the water was kept at 70 to 75 degrees Fahr. The oil used was a straight-run mineral oil having a viscosity of 100 Saybolt seconds at 100 degrees Fahr. The temperature of the oil was kept at 108 to 115 degrees Fahr. No appreciable rise in temperature was obtained during quenching.

*Heating for Quenching*—Specimens were heated in an electric muffle furnace. All specimens were supported at the ends on racks while heating. No attempt was made to prevent scaling, the intention being to heat the specimens under conditions existing in usual commercial practice. Three or four specimens of the ½ and 1-inch section sizes were heated at a time, while only two specimens of the 2 and 3-inch section sizes were heated at a time. Since each specimen was quenched separately, the pieces were actually held at temperature from 30 to 40 minutes. The specimens were introduced when the furnace was at temperature.

The progress of heating and the time at temperature were followed by observing the temperature indicated with a thermocouple inserted in the center of a dummy specimen of the same size as the specimens being treated. The center of the 3-inch bars reached temperature in approximately 55 minutes.

The temperature distribution within the furnace was explored and found to vary no more than 5 degrees Fahr. within the area occupied by the specimens. The temperature fluctuation due to the lag of the control mechanism was ± 10 degrees Fahr.

*Quenching Operation*—After the specimens had been at temperature for a minimum of 30 minutes, they were quenched in a

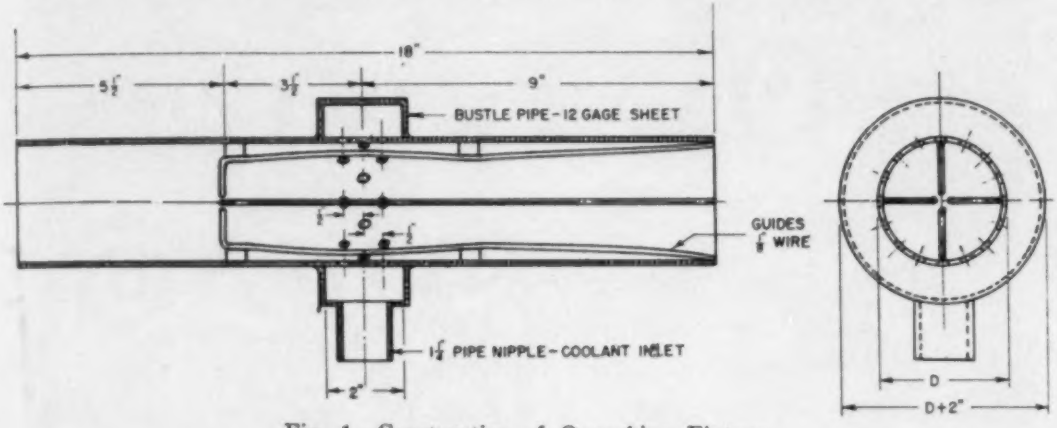


Fig. 1—Construction of Quenching Fixture.

fixture in running water or oil, depending on the coolant used. The time elapsed during the transfer of the specimen from the furnace into the quenching fixture and until it was immersed in the coolant was 4 to 5 seconds or less. All specimens were kept under the circulating coolant until they could be handled with bare hands.

## Description of Quenching Fixture

In order to obtain uniform results on quenching, the temperature and velocity of the coolant against the surface of the specimen must be held constant. After a consideration of several possible methods of quenching involving the control of these two factors, the method finally adopted was that described in the following paragraphs.

In brief, the quenching is effected in a tube or fixture suspended horizontally above the quenching bath until the specimen is inserted. Immediately after the specimen is inserted in the fixture, the fixture and specimen drop into the bath. Coolant enters the fixture through a series of twenty-six ¼-inch circumferential holes located under a bustle pipe attached to the center of the tube, impinges upon the specimen at its center, and then divides to discharge from the fixture at both ends. The coolant automatically enters the

fixture just as the latter is submerged in the bath. A sketch of the specimen fixture or tube is shown in Fig. 1.

The tubes used in constructing the fixtures for the four bar sizes were chosen of such internal diameter that the annular area between the inside of the tube and the specimen was 5.5 square inches. This gave assurance that the velocity of the water or oil in the fixture would be the same with the pump delivering a constant volume of coolant. The tubes were 18 inches long.

The bustle pipes were 2 inches wide and 1 inch deep and were torch-welded to the tubes. A 1¼-inch threaded nipple welded to each bustle pipe was used to attach the fixtures to the circulating system by having it screw into a 1¼-inch 3-way brass stopcock connected to the supply line from the pump. Guides inside the fixtures were necessary for facilitating the entrance of specimens and for centrally locating the specimens in the tubes. These guides consisted of ⅛-inch wires welded to two narrow supports made of angle iron set about 6½ inches apart. The ends of the guides toward the entrance through which the specimens are introduced in the fixture were bent outward and welded to the inside of the tube. The other ends were bent sharply inward to 90 degrees to serve as a stop for locating the specimen centrally with regard to the two ends of the tube. The guide wires were also bent outward between the two supports so that the specimens rested at only two points on a guide. The dimensions of tubes for the fixtures and the distance of the guides from the inside wall of the tubes are listed in Table II.

**Table II**
**Dimensions of Tubes and Height of Guides for Quenching Fixtures**

| Diameter of Specimen Inches | Pipe Size | | No. of Guides | Height of Guides Inches |
|---|---|---|---|---|
| | O.D. | I.D. | | |
| ½ | 2⅞ | 2.685 | 6 | 1.04 |
| 1 | 3⅛ | 2.813 | 6 | 0.85 |
| 2 | 3½ | 3.310 | 4 | 0.60 |
| 3 | 4¼ | 4.010 | 4 | 0.45 |

During quenching a flow of 60 gallons of coolant per minute was maintained. With both ends of the quenching fixture opened and with a specimen in the fixture, the coolant velocity within the fixture was 105 feet per minute. The amount of coolant delivered by the pump through a given size orifice was the same whether oil or water was employed.

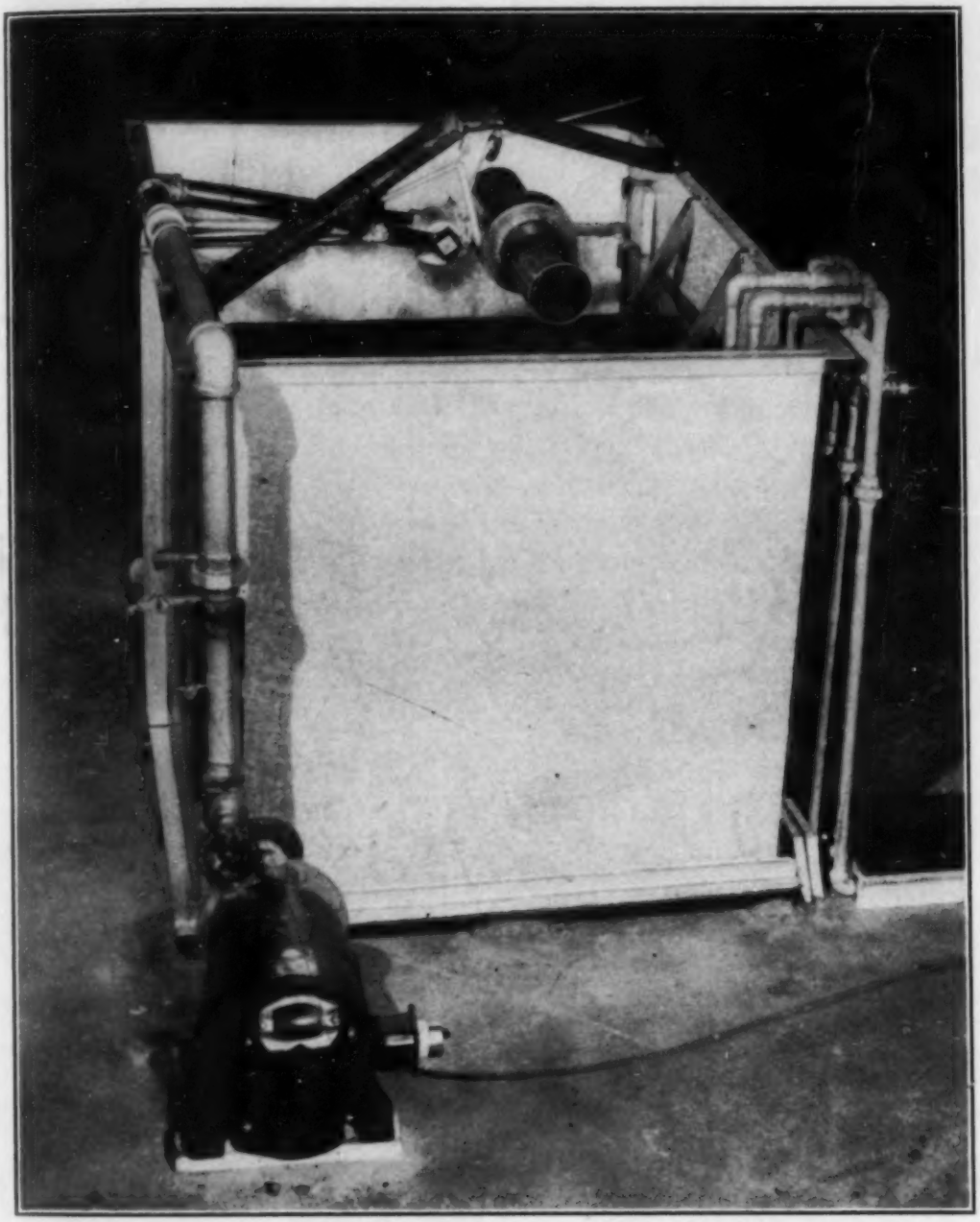

Fig. 2—Quenching Apparatus with One of the Fixtures Outside of Bath in Position Ready to Insert the Specimen.

A photograph of the quenching apparatus is shown in Fig. 2. The tank used was large enough to hold 220 gallons of oil or water. Copper tubing of ¾ inch O.D. installed for heating or cooling the bath was arranged along the sides of the tank. A centrifugal pump, attached to the tank, was used to circulate the coolant.

*Operation of Quenching Apparatus*—In operating the quench-

ing apparatus the temperature of the coolant was first brought within the proper range either by cooling or by heating. Just prior to the removal of the specimen from the furnace the quenching fixture was suspended above the bath, the three-way stopcock was set to discharge the coolant into the bath rather than into the fixture and the circulation of the coolant was started. A simple foot arrangement was designed to trip the fixture so that it could swing downward about a pipe elbow attached loosely to the supply line from the pump. A spring attached to an arm on the three-way stopcock automatically closed the port discharging the coolant into the bath and opened a port into the quenching fixture delivering the coolant to the bustle pipe a fraction of a second after the fixture was tripped. The successful operation of the fixture depended on keeping the port of the stopcock leading into the bustle pipe closed until the fixture was submerged, otherwise nonuniform hardening occurred. From 4 to 5 seconds were normally consumed in removing the specimen from the furnace and quenching it.

## Uniformity of Quench

*Oil Quenching*—S.A.E. 4340 steel, in all section sizes, was taken for trials to test the uniformity in hardness obtained with the quenching fixture. After quenching, each bar was cut into 1-inch slices and prepared for hardness surveys. Transverse hardness readings were taken on two diameters for each inch of length of the 7-inch long specimens. The results are summarized in Table III. These data show that the end slices were somewhat softer than the middle slices. There was, however, a very close agreement in the hardness of the middle three 1-inch slices.

Surface hardness values taken only on the 1, 2, and 3-inch rounds are shown in Table IV, and are reported both as average values and as the mode* of all readings. The surface hardness tended to decrease toward the ends of the specimens, but the decrease was not pronounced.

The results of the survey indicated that good uniformity in hardness could be obtained. There was no evidence of off-center hardening.

*Water Quenching*—A similar hardness survey was not conducted on specimens of water-hardening steels. Visual examination

*Mode value refers to that hardness value in a series of hardness values that occurs most frequently.

**Table III**

**Transverse Hardness Values on 1-Inch Slices of S.A.E. 4340 Steel. To Show Variations in Hardness in Longitudinal Direction Along 7-Inch Specimens**

| | Position of 1-Inch Slices. Distance from One End | | | | | |
|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 5 | 6 | 7 |
| | | | ½-In. Diameter Bar | | | | |
| 1 | 52.2 | 54.7 | 55.0 | 55.0 | 55.0 | 54.2 | 54.0 |
| 2 | 52.7 | 54.7 | 56.2 | 55.2 | 55.2 | 54.7 | 54.0 |
| 3 | 52.7 | 54.7 | 56.0 | 55.7 | 55.2 | 55.0 | 54.0 |
| C | 52.0 | 55.0 | 56.0 | 55.0 | 55.0 | 54.0 | 54.0 |
| | | | 1-In. Diameter Bar | | | | |
| 1 | 53.5 | 53.0 | 55.8 | 56.0 | 55.0 | 54.3 | 52.8 |
| 2 | 53.5 | 53.0 | 55.8 | 55.5 | 54.8 | 55.0 | 52.8 |
| 3 | 53.8 | 52.5 | 55.8 | 55.5 | 55.3 | 55.0 | 53.0 |
| 4 | 53.8 | 53.2 | 56.5 | 57.0 | 55.3 | 54.8 | 53.3 |
| 5 | 53.8 | 53.5 | 56.5 | 56.7 | 56.0 | 55.0 | 53.5 |
| 6 | 53.8 | 53.0 | 56.5 | 56.5 | 56.3 | 55.5 | 53.5 |
| 7 | 53.2 | 53.0 | 56.2 | 56.3 | 56.0 | 55.0 | 52.8 |
| C | 52.0 | 53.0 | 56.0 | 56.2 | 57.0 | 54.8 | 52.6 |
| | | | 2-In. Diameter Bar | | | | |
| 1 | .... | 53.0 | 54.0 | 54.0 | 53.0 | 51.8 | .... |
| 2 | .... | 53.0 | 54.0 | 53.5 | 53.0 | 51.8 | .... |
| 3 | .... | 53.0 | 53.8 | 53.8 | 52.8 | 52.3 | .... |
| 4 | .... | 53.0 | 53.5 | 53.5 | 52.8 | 52.5 | .... |
| 5 | .... | 53.3 | 53.5 | 53.5 | 52.8 | 52.0 | .... |
| 6 | .... | 53.0 | 53.5 | 53.3 | 52.8 | 52.3 | .... |
| 7 | .... | 53.0 | 53.3 | 53.3 | 52.8 | 52.5 | .... |
| 8 | .... | 53.0 | 53.3 | 53.5 | 52.8 | 52.5 | .... |
| 9 | .... | 53.0 | 54.0 | 53.3 | 52.5 | 51.8 | .... |
| 10 | .... | 53.5 | 53.5 | 53.5 | 52.8 | 52.3 | .... |
| 11 | .... | 53.5 | 53.5 | 53.3 | 52.8 | 52.3 | .... |
| 12 | .... | 53.3 | 53.5 | 53.3 | 53.0 | 51.8 | .... |
| 13 | .... | 53.8 | 54.3 | 52.8 | 52.3 | 51.5 | .... |
| 14 | .... | 53.3 | 53.5 | 52.5 | 51.5 | 51.5 | .... |
| 15 | .... | 53.0 | 53.3 | 53.3 | 51.8 | 52.0 | .... |
| C | .... | 53.0 | 53.2 | 52.2 | 51.6 | 52.0 | .... |
| | | | 3-In. Diameter Bar | | | | |
| 1 | .... | 51.8 | 53.5 | 53.0 | 51.8 | 51.8 | .... |
| 2 | .... | 51.8 | 53.3 | 53.0 | 51.8 | 51.5 | .... |
| 3 | .... | 51.8 | 53.3 | 53.0 | 51.5 | 51.5 | .... |
| 4 | .... | 52.0 | 53.0 | 53.0 | 51.0 | 51.3 | .... |
| 5 | .... | 51.8 | 52.5 | 52.8 | 50.5 | 50.7 | .... |
| 6 | .... | 51.5 | 52.3 | 52.0 | 50.5 | 51.5 | .... |
| 7 | .... | 51.5 | 52.3 | 52.0 | 50.8 | 51.3 | .... |
| 8 | .... | 51.5 | 52.0 | 51.8 | 51.0 | 51.3 | .... |
| 9 | .... | 51.3 | 51.8 | 51.8 | 50.3 | 51.5 | .... |
| 10 | .... | 50.8 | 50.5 | 51.3 | 50.0 | 50.8 | .... |
| 11 | .... | 50.8 | 50.8 | 51.5 | 50.5 | 50.8 | .... |
| 12 | .... | 50.8 | 51.3 | 51.5 | 49.8 | 50.3 | .... |
| 13 | .... | 50.8 | 50.8 | 51.3 | 50.0 | 51.0 | .... |
| 14 | .... | 50.5 | 50.5 | 50.8 | 49.5 | 51.3 | .... |
| 15 | .... | 50.8 | 50.5 | 50.5 | 50.0 | 50.8 | .... |
| 16 | .... | 51.0 | 49.8 | 50.5 | 50.0 | 50.5 | .... |
| 17 | .... | 50.3 | 49.3 | 50.3 | 50.0 | 50.5 | .... |
| 18 | .... | 50.0 | 49.5 | 49.3 | 49.5 | 49.5 | .... |
| 19 | .... | 48.5 | 49.3 | 49.3 | 48.5 | 48.8 | .... |
| 20 | .... | 48.8 | 48.0 | 48.3 | 48.3 | 47.5 | .... |
| 21 | .... | 49.5 | 47.3 | 48.5 | 47.5 | 47.5 | .... |
| 22 | .... | 46.8 | 45.8 | 46.3 | 46.3 | 46.8 | .... |
| 23 | .... | 46.3 | 46.8 | 45.8 | 45.0 | 46.8 | .... |
| C | .... | 46.0 | 46.5 | 45.8 | 45.2 | 47.0 | .... |

Note—Slices were cut from quenched specimens, 7 inches long and marked in consecutive order starting from one end.

All columns of readings start with the hardness $\frac{1}{16}$ inch from surface and proceed downward to the hardness at the center. Readings taken at $\frac{1}{16}$ inch intervals along a radius.

of sections of two of the steels appeared to serve the same purpose. Photographs of etched sections of water-quenched 3-inch rounds of S.A.E. 1045 fine-grained steel and S.A.E. 3130 steel are shown in

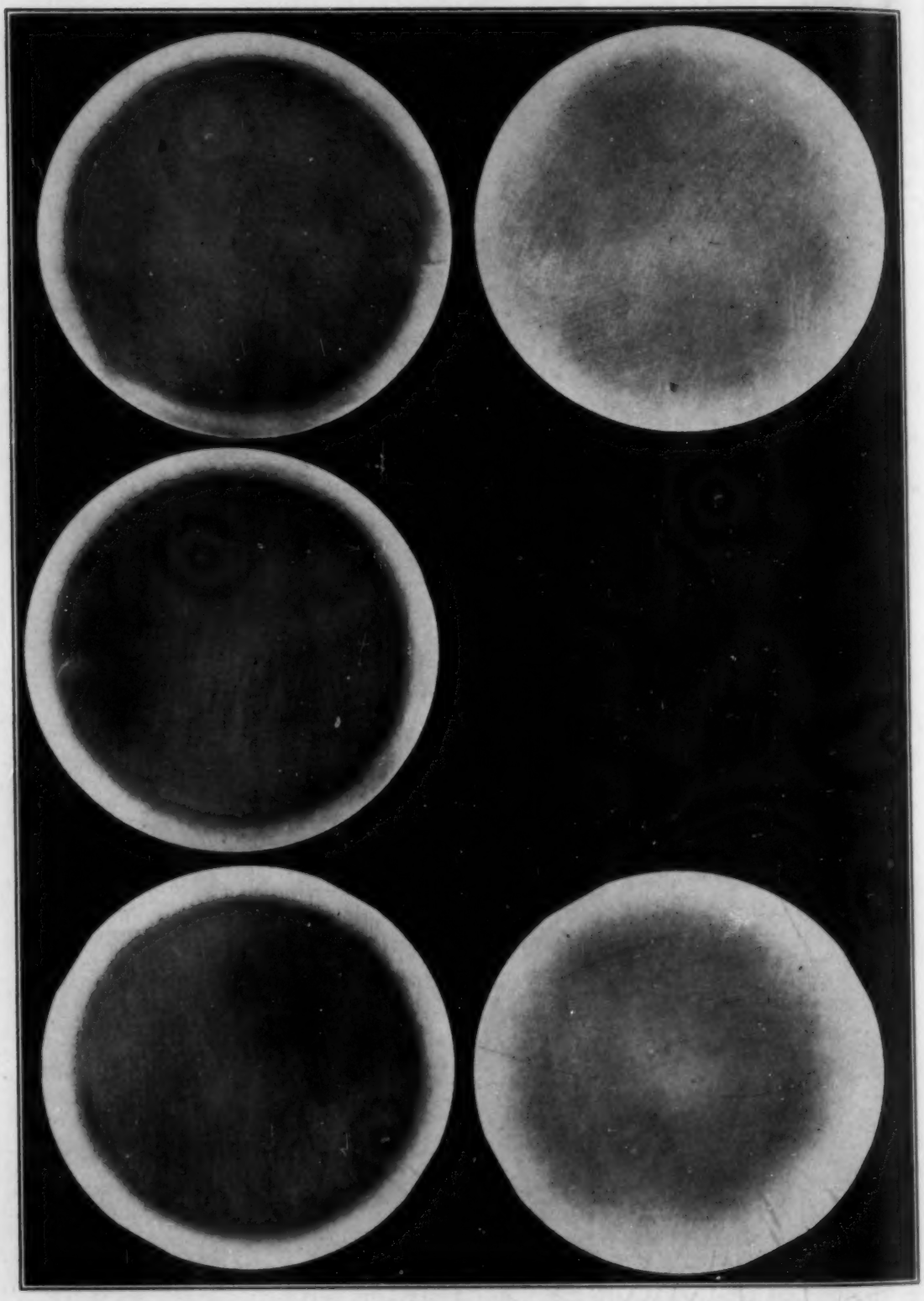

Fig. 3—Etched Sections of Water-Quenched 3-Inch Rounds of S.A.E. 1045 Fine-Grained Steel and S.A.E. 3130 Steel. Left—S.A.E. 1045, Fine-Grained. Bottom to Top—4, 2, and 1 Inch from One End of 7-Inch Long Bar. Right—S.A.E. 3130. Bottom to Top—4 and 1 Inch from One End of 7-Inch Long Bar.

**Table IV**

**Surface Hardness Values on 1-Inch Slices of S.A.E. 4340 Steel to Show Variations Obtained After Quenching**

| Length Intervals Along Bar | Rockwell "C" Hardness | | | | | | | | Mode of Readings | Average |
|---|---|---|---|---|---|---|---|---|---|---|
| | 57 | 56 | 55 | 54 | 53 | 52 | 51 | 50 | | |
| **1-Inch Diameter Bar** | | | | | | | | | | |
| End—1" | .. | .. | 6 | 12 | 13 | 6 | 3 | .. | 53 | 53.3 |
| 1—2" | .. | .. | 15 | 23 | 2 | .. | .. | .. | 54 | 54.3 |
| 2—3" | .. | 8 | 10 | 17 | 5 | .. | .. | .. | 54 | 54.5 |
| 3—4" | 6 | 11 | 12 | 10 | 1 | .. | .. | .. | 55 | 55.3 |
| 4—5" | 9 | 21 | 7 | 3 | .. | .. | .. | .. | 56 | 56.1 |
| 5—6" | .. | .. | 6 | 22 | 12 | .. | .. | .. | 54 | 54.2 |
| 6—7" | .. | 10 | 8 | 11 | 1 | .. | .. | .. | 54 | 54.1 |
| **2-Inch Diameter Bar** | | | | | | | | | | |
| 2—3" | 3 | 10 | 12 | 9 | 6 | .. | .. | .. | 55 | 55.1 |
| 3—4" | .. | 7 | 13 | 12 | 6 | 2 | .. | .. | 55 | 54.5 |
| 4—5" | .. | 1 | 11 | 15 | 13 | .. | .. | .. | 54 | 54.0 |
| 5—6" | .. | .. | 6 | 8 | 26 | .. | .. | .. | 53 | 53.5 |
| 6—7" | .. | .. | 8 | 18 | 13 | 1 | .. | .. | 54 | 54.2 |
| **3-Inch Diameter Bar** | | | | | | | | | | |
| 2—3" | .. | .. | 6 | 10 | 9 | 15 | .. | .. | 52 | 53.2 |
| 3—4" | 4 | 9 | 11 | 12 | 4 | .. | .. | .. | 54 | 54.9 |
| 4—5" | .. | 1 | 4 | 5 | 13 | 6 | 9 | 2 | 53 | 52.7 |
| 5—6" | .. | .. | .. | 6 | 19 | 10 | 5 | .. | 53 | 52.7 |
| 6—7" | .. | .. | .. | 5 | 19 | 11 | 5 | .. | 53 | 52.7 |

Note: Slices were cut from quenched specimens 7 inches long and marked in consecutive order starting from one end.

Fig. 3 to illustrate the absence of off-center hardness in the middle sections of the specimens and the uniformity in hardness in the longitudinal direction.

## Method of Conducting the Hardenability Studies

*Hardness Test Specimens*—Two inches from each end of the quenched specimens was discarded. The remaining 3-inch pieces were cut into three 1-inch slices. The middle 1-inch slice was used in exploring the transverse hardness of the steels in the quenched condition. All three slices were later used in determining the hardness of the quenched steels after tempering.

*Cutting*—A Manhattan No. 415 abrasive, rubber-bonded, cut-off wheel $\frac{1}{16}$ inch in thickness and 14 inches in diameter was used to slice the quenched bars. Considerable trouble was encountered in obtaining a wheel which would reduce overheating to a minimum and over a dozen wheels from several manufacturers were tried before a satisfactory one was found.

*Preparation of Surfaces for Testing*—A minimum of 0.030 inch was ground from each of the cut surfaces. The ground surfaces were then polished through 0 paper and etched with 5 per cent nital solution to detect overheating regions of off-center hardening before making the hardness tests. The same procedure was fol-

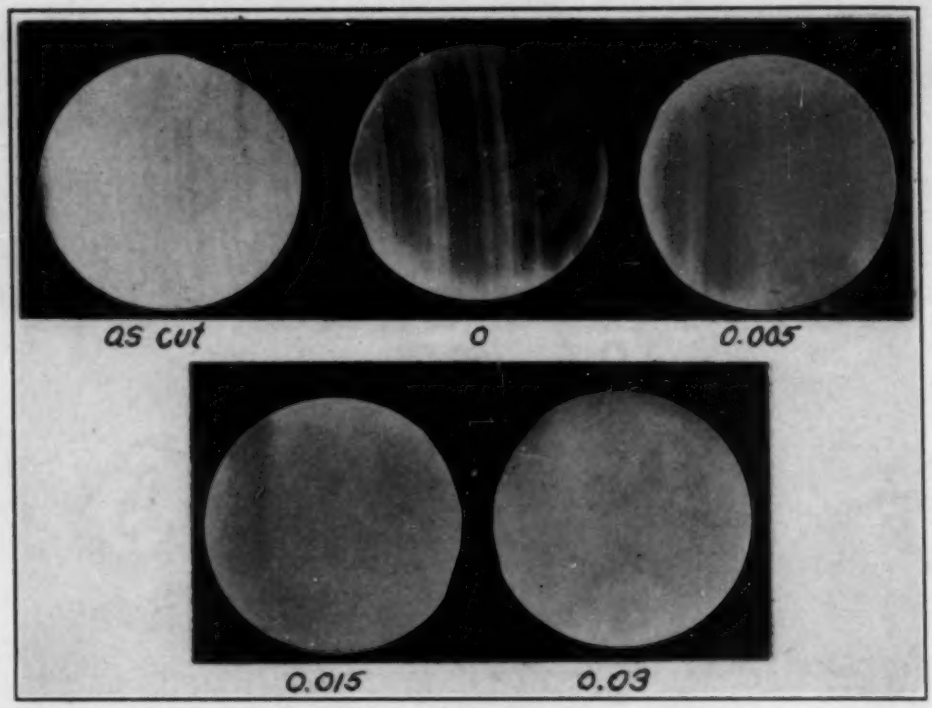


Fig. 4—Depth of Tempering Obtained After Sectioning a Quenched Bar with an Abrasive Cut-Off Wheel. S.A.E. 2345, 1-Inch Diameter. Etched in 5 Per Cent Nital. Amount of Metal Removed from Surface After Cutting Off is Indicated by the Set of Figures on the Photograph.

lowed in preparing the surfaces of sections to be tempered. Fig. 4 shows the unetched surface after the cut-off and the etched surfaces following the removal of various amounts of metal with coarse emery paper.

*Transverse Hardness*—Transverse Rockwell "C" hardness values were obtained on the middle 1-inch slice, the readings being taken every sixteenth of an inch on two diameters at right angles to each other. A specially designed fixture similar to that described by Bain and Davenport (3)[1] was used to locate the points for the impressions and to support the specimens on the hardness testing instrument.

*Surface Hardness*—To determine the surface hardness of each specimen four longitudinal flats, 90 degrees apart, were prepared by grinding and polishing approximately 0.010 to 0.015 inch from the surface. Ten readings were taken on each flat making a total of 40 readings for a specimen. The surface hardness was taken as the mode of the 40 readings rather than as the average value. Both values usually fell within a point of each other.

*Center Hardness*—The value taken for center hardness was the

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

average of five readings, consisting of the center reading and four readings on a circle $\frac{1}{16}$ inch away from center. This method of reporting center hardnesses was suggested by Burns, Moore and Archer (1).

## Tempering Tests

*Heating for Tempering*—As previously stated, each test piece was polished and etched to insure elimination of overheated regions before proceeding with the tempering treatments. Two test pieces of each steel were used of each section size for any one temperature. They were tempered in different furnace charges in order to check the practice. The tempering operation was conducted in an electric furnace provided with a fan to circulate the air around the specimens.

*Tempering Temperatures*—The oil-quenched specimens were tempered at 400, 800, and 1000 degrees Fahr. The water-quenched pieces were tempered at 800, 1000, and 1200 degrees Fahr. All specimens were held 1 hour at temperature, and were then removed from the furnace and cooled in air. Time at temperature was obtained by observing the temperature of a dummy test piece having a thermocouple inserted at the center.

*Hardness Testing of Tempered Specimens*—Hardness values were obtained after removing the scale and polishing the surfaces through No. 0 paper. Rockwell "C" readings were first obtained. They were followed by Brinell readings. Readings were taken at the center, at half radius, and on the surface of each specimen.

The Rockwell reading at the center was the average of five values (one at the center and four $\frac{1}{16}$ inch from center); that at half radius was the average of four values; and that at the surface was the mode of forty values.

A single Brinell reading was taken at the center, four readings were taken at half radius and four at the surface, one value on each flat. The Brinell impressions on all oil-quenched steels, except S.A.E. steels 4145, 5145, and 6145, were made with a standard 10-mm. steel ball with a 3000-kilogram load applied for 30 seconds. When the Brinell values exceeded 500, the steel ball was checked frequently for permanent deformation. The necessity for frequent checking and replacement of steel balls when the hardness exceeded 500 Brinell was later eliminated by employing a tungsten carbide ball. This ball was used on all the water-quenched steels and on the S.A.E. steels 4145, 5145, and 6145 of the oil-hardening group. According

**Table V**

**Rockwell "C" Hardness Values for Water-Quenched S.A.E. 4130 Steel in ½, 1, 2 and 3-Inch Round Sections, Including Penetration and Surface Readings**

All section sizes were quenched in water as 7-inch long specimens; 2 inches from each end was discarded, the middle 3-inch piece was cut into three 1-inch slices and the middle slice was used for hardenability tests

½-Inch Round

| | Specimen A | | | | Specimen B | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 51 | 51 | 52 | 52 | 52 | 51 | 52 | 52 | 51.6 |
| 2 | 51 | 51 | 52 | 53 | 51 | 51 | 52 | 52 | 51.6 |
| 3 | 52 | 53 | 53 | 53 | 52 | 52 | 53 | 53 | 52.6 |
| C | 53 | .. | .. | .. | 53 | .. | .. | .. | 52.7 |

1-Inch Round

| | Specimen A | | | | Specimen B | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 51 | 51 | 51 | 52 | 52 | 52 | 52 | 52 | 51.6 |
| 2 | 51 | 51 | 51 | 51 | 52 | 52 | 52 | 51 | 51.4 |
| 3 | 50 | 50 | 51 | 50 | 52 | 52 | 50 | 50 | 50.6 |
| 4 | 49 | 50 | 50 | 49 | 51 | 51 | 50 | 49 | 49.9 |
| 5 | 48 | 49 | 49 | 48 | 50 | 51 | 49 | 49 | 49.1 |
| 6 | 47 | 48 | 48 | 48 | 49 | 49 | 49 | 48 | 48.3 |
| 7 | 46 | 47 | 48 | 47 | 48 | 48 | 48 | 48 | 47.5 |
| C | 47 | .. | .. | .. | 48 | .. | .. | .. | 47.5 |

Surface Hardness Frequency

| Section Size | 56 | 55 | 54 | 53 | 52 | 51 | |
|---|---|---|---|---|---|---|---|
| ½ inch A | .. | 8 | 11 | 4 | 9 | 8 | |
| ½ inch B | 4 | 20 | 7 | 8 | 1 | .. | |
| Surface hardness of ½-inch round | | | | | | | 55 |
| 1 inch A | 1 | 33 | 6 | | | | |
| 1 inch B | 14 | 25 | 1 | | | | |
| Surface hardness of 1-inch round | | | | | | | 55 |
| 2 inch A | | 4 | 23 | 11 | 2 | | |
| 2 inch B | | 9 | 25 | 3 | 3 | | |
| Surface hardness of 2-inch round | | | | | | | 54 |
| 3 inch A | | 3 | 17 | 18 | 2 | | |
| 3 inch B | | 15 | 23 | 2 | .. | | |
| Surface hardness of 3-inch round | | | | | | | 54 |

2-Inch Round

| | Specimen A | | | | Specimen B | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 51 | 51 | 51 | 51 | 52 | 51 | 52 | 52 | 51.4 |
| 2 | 50 | 50 | 51 | 50 | 52 | 50 | 51 | 51 | 50.6 |
| 3 | 48 | 49 | 49 | 48 | 50 | 48 | 50 | 49 | 48.9 |
| 4 | 45 | 46 | 47 | 46 | 48 | 47 | 48 | 48 | 46.9 |
| 5 | 43 | 44 | 44 | 43 | 46 | 44 | 46 | 44 | 44.3 |
| 6 | 41 | 42 | 42 | 41 | 43 | 42 | 43 | 42 | 42.0 |
| 7 | 40 | 40 | 40 | 40 | 41 | 40 | 41 | 40 | 40.3 |
| 8 | 38 | 38 | 38 | 38 | 39 | 38 | 39 | 38 | 38.3 |
| 9 | 37 | 36 | 37 | 36 | 38 | 36 | 38 | 36 | 36.8 |
| 10 | 37 | 34 | 36 | 36 | 35 | 34 | 34 | 34 | 35.0 |
| 11 | 34 | 34 | 35 | 35 | 34 | 35 | 35 | 34 | 34.5 |
| 12 | 34 | 34 | 34 | 35 | 34 | 34 | 34 | 33 | 34.0 |
| 13 | 32 | 33 | 34 | 34 | 33 | 34 | 34 | 34 | 33.5 |
| 14 | 33 | 33 | 34 | 34 | 34 | 33 | 34 | 34 | 33.7 |
| 15 | 34 | 35 | 35 | 34 | 33 | 32 | 34 | 35 | 34.0 |
| C | 33 | .. | .. | .. | 34 | .. | .. | .. | 33.9 |

to Petrenko and co-workers (4), the tungsten carbide ball gives Brinell numbers above 500 that are 15 points or more above the values obtained with a steel ball. In other words, the tungsten carbide ball gives truer hardness values in this range of hardness. This should be taken into account in considering the Brinell values of the very hard tempered specimens read with the two types of balls.

**Table V (Continued)**

3-Inch Round

| | Specimen A | | | | Specimen B | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 51 | 49 | 52 | 51 | 52 | 51 | 50 | 50 | 50.8 |
| 2 | 51 | 49 | 51 | 50 | 51 | 48 | 48 | 48 | 49.5 |
| 3 | 48 | 46 | 49 | 46 | 49 | 45 | 46 | 45 | 46.8 |
| 4 | 45 | 44 | 46 | 44 | 47 | 45 | 44 | 42 | 44.6 |
| 5 | 43 | 42 | 43 | 41 | 43 | 41 | 42 | 40 | 41.9 |
| 6 | 39 | 38 | 40 | 38 | 40 | 39 | 40 | 37 | 38.9 |
| 7 | 35 | 35 | 37 | 35 | 37 | 34 | 37 | 34 | 35.5 |
| 8 | 34 | 33 | 34 | 33 | 35 | 33 | 35 | 33 | 33.8 |
| 9 | 33 | 32 | 34 | 32 | 33 | 32 | 33 | 32 | 32.6 |
| 10 | 32 | 32 | 32 | 31 | 31 | 30 | 33 | 30 | 31.4 |
| 11 | 30 | 30 | 31 | 29 | 31 | 31 | 31 | 30 | 30.4 |
| 12 | 30 | 29 | 31 | 29 | 30 | 29 | 30 | 30 | 29.8 |
| 13 | 30 | 29 | 31 | 27 | 30 | 29 | 31 | 29 | 29.5 |
| 14 | 30 | 29 | 29 | 28 | 30 | 30 | 28 | 29 | 29.1 |
| 15 | 29 | 28 | 29 | 27 | 30 | 30 | 28 | 29 | 28.8 |
| 16 | 29 | 28 | 28 | 28 | 29 | 29 | 28 | 28 | 28.4 |
| 17 | 29 | 27 | 29 | 27 | 29 | 29 | 28 | 28 | 28.3 |
| 18 | 28 | 28 | 29 | 28 | 30 | 28 | 27 | 28 | 28.3 |
| 19 | 28 | 28 | 29 | 29 | 30 | 29 | 28 | 29 | 28.8 |
| 20 | 26 | 28 | 30 | 28 | 31 | 30 | 27 | 29 | 28.6 |
| 21 | 28 | 27 | 31 | 28 | 29 | 31 | 28 | 30 | 29.0 |
| 22 | 27. | 29 | 31 | 28 | 30 | 31 | 28 | 31 | 29.4 |
| 23 | 29 | 29 | 30 | 29 | 29 | 30 | 29 | 31 | 29.5 |
| C | 28 | .. | .. | .. | 29 | .. | .. | .. | 29.3 |

All columns of penetration readings start with the hardness $\frac{1}{32}$ inch from the surface and proceed downward to the center. Readings taken every $\frac{1}{16}$ of an inch along a radius. Average center hardness is the average of center and values on a circle $\frac{1}{16}$ inch away from center.

## Presentation of Data

Data for each of the water-quenched steels are plotted in Figs. 5 to 12 inclusive. Table V is a sample tabulation of the hardness values including the penetration and surface readings, recorded for each of the water-quenched steels. The tabulation of the tempering data for the water-quenched steels is illustrated in Table VI.

Similar data for the oil-hardened steels are plotted in Figs. 13 to 20 inclusive. Sample data sheets for the oil-quenched steels are shown in Table VII and VIII.

The conventional hardness penetration curves appear on the left side of the figures while the tempering curves appear to the right. The tempering data were grouped and plotted according to section size. The upper set of curves shows the changes in hardness, for different tempering temperatures, at the surface, at half radius and at the center for the 1, 2, and 3-inch diameter bars. The lower set of curves shows the change in hardness across the section for each tempering temperature. Brinell values are indicated by the full lines, while Rockwell "C" values are indicated by the dashed lines.

**Table VI**

**Brinell and Rockwell C Hardness Values of Water-Quenched S.A.E. 4130 Steel**

| Section Size | Rockwell "C" Hardness Values | | | | | | | | | | | Brinell Hardness Values* | | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | Center Values | | | | | | Values for ½ Distance Between Center and Edge | | | | | Mode of Surface Frequency | Center Values | Values for ½ Distance Between Center and Edge | | | | | Surface Values | | | | |
| | Center | 1 | 2 | 3 | 4 | Avg. | 1 | 2 | 3 | 4 | Avg. | | | 1 | 2 | 3 | 4 | Avg. | 1 | 2 | 3 | 4 | Avg. |
| | Tempered at 800 Degrees Fahr. | | | | | | | | | | | | | | | | | | | | | | |
| 1 inch A | 35 | 40 | 40 | 40 | 41 | 39.2 | 41 | 41 | 42 | 42 | 41.5 | 41 | 388 | 401 | 406 | 409 | 406 | 406 | 415 | 415 | 409 | 415 | 413 |
| 1 inch B | 35 | 39 | 39 | 39 | 40 | 38.4 | 41 | 41 | 41 | 42 | 41.3 | 42 | 388 | 401 | 401 | 393 | 393 | 397 | 409 | 406 | 409 | 409 | 408 |
| 2 inch A | 29 | 32 | 32 | 32 | 34 | 31.8 | 34 | 33 | 34 | 34 | 33.8 | 40 | 311 | 341 | 337 | 341 | 341 | 340 | 401 | 409 | 415 | 415 | 410 |
| 2 inch B | 29 | 32 | 32 | 32 | 32 | 31.4 | 35 | 35 | 34 | 34 | 34.5 | 41 | 311 | 341 | 341 | 341 | 341 | 341 | 415 | 401 | 409 | 409 | 408 |
| 3 inch A | 25 | 27 | 27 | 27 | 28 | 26.8 | 29 | 28 | 28 | 27 | 28.0 | 40 | 285 | 288 | 298 | 293 | 293 | 293 | 401 | 409 | 409 | 401 | 405 |
| 3 inch B | 26 | 28 | 27 | 28 | 28 | 27.4 | 29 | 28 | 27 | 27 | 27.8 | 40 | 285 | 293 | 293 | 293 | 293 | 293 | 406 | 409 | 401 | 409 | 406 |
| | Tempered at 1000 Degrees Fahr. | | | | | | | | | | | | | | | | | | | | | | |
| 1 inch A | 31 | 33 | 33 | 34 | 34 | 33.0 | 35 | 34 | 34 | 34 | 34.3 | 36 | 331 | 341 | 341 | 341 | 341 | 341 | 341 | 341 | 341 | 341 | 341 |
| 1 inch B | 32 | 33 | 33 | 33 | 33 | 32.8 | 35 | 34 | 34 | 34 | 34.3 | 35 | 321 | 337 | 331 | 337 | 331 | 334 | 341 | 341 | 341 | 341 | 341 |
| 2 inch A | 26 | 27 | 26 | 26 | 27 | 26.4 | 29 | 28 | 28 | 27 | 28.0 | 34 | 277 | 285 | 285 | 293 | 288 | 288 | 337 | 335 | 341 | 341 | 339 |
| 2 inch B | 26 | 27 | 27 | 27 | 27 | 26.8 | 28 | 29 | 29 | 29 | 28.8 | 34 | 277 | 293 | 293 | 297 | 293 | 294 | 337 | 331 | 335 | 341 | 336 |
| 3 inch A | 23 | 25 | 25 | 23 | 25 | 24.2 | 24 | 23 | 24 | 23 | 23.5 | 34 | 255 | 266 | 269 | 269 | 262 | 267 | 341 | 335 | 337 | 341 | 339 |
| 3 inch B | 23 | 24 | 25 | 23 | 25 | 24.0 | 25 | 26 | 24 | 25 | 25.0 | 34 | 262 | 269 | 269 | 269 | 269 | 269 | 345 | 341 | 345 | 345 | 344 |
| | Tempered at 1200 Degrees Fahr. | | | | | | | | | | | | | | | | | | | | | | |
| 1 inch A | 25 | 26 | 27 | 25 | 26 | 25.8 | 25 | 25 | 26 | 25 | 25.3 | 25 | 269 | 269 | 269 | 269 | 266 | 268 | 269 | 272 | 272 | 272 | 271 |
| 1 inch B | 24 | 25 | 25 | 24 | 25 | 24.6 | 26 | 25 | 24 | 26 | 25.3 | 27 | 262 | 269 | 269 | 266 | 262 | 267 | 269 | 269 | 269 | 269 | 269 |
| 2 inch A | 20 | 19 | 19 | 19 | 21 | 19.6 | 21 | 20 | 21 | 21 | 20.8 | 25 | 229 | 241 | 241 | 241 | 241 | 241 | 269 | 269 | 269 | 269 | 269 |
| 2 inch B | 18 | 19 | 18 | 18 | 19 | 18.4 | 20 | 20 | 19 | 21 | 20.0 | 24 | 231 | 241 | 241 | 241 | 241 | 241 | 262 | 262 | 269 | 269 | 266 |
| 3 inch A | 16 | 17 | 17 | 17 | 17 | 16.8 | 16 | 17 | 17 | 17 | 16.8 | 23 | 217 | 229 | 225 | 226 | 223 | 226 | 269 | 269 | 269 | 269 | 269 |
| 3 inch B | 14 | 16 | 16 | 17 | 17 | 16.0 | 17 | 17 | 15 | 18 | 16.8 | 25 | 217 | 226 | 223 | 229 | 229 | 227 | 269 | 269 | 262 | 269 | 267 |

*Tungsten carbide ball.

**Table VII**

**Rockwell C Hardness Values for Oil-Quenched S.A.E. 6145 Steel in ½, 1, 2 and 3-Inch Round Sections, Including Penetration and Surface Readings**

All section sizes were quenched in oil as specimens 7 inches long. 2 inches from each end was discarded, the middle 3-inch piece was cut into three 1-inch slices and the middle slice was used for hardenability tests.

Hardness Penetrations

½-Inch Round

| | Specimen C | | | | Specimen D | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 61 | 61 | 61 | 61 | 61 | 61 | 60 | 61 | 60.9 |
| 2 | 61 | 61 | 61 | 62 | 61 | 61 | 61 | 61 | 61.1 |
| 3 | 61 | 61 | 62 | 62 | 61 | 61 | 61 | 62 | 61.4 |
| C | 62 | .. | .. | .. | 62 | .. | .. | .. | 61.5 |

1-Inch Round

| | Specimen | | | | Specimen | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 61 | 60 | 60 | 60 | 60 | 60 | 60 | 60 | 60.1 |
| 2 | 60 | 60 | 60 | 60 | 60 | 59 | 60 | 60 | 59.9 |
| 3 | 59 | 59 | 60 | 59 | 59 | 58 | 59 | 60 | 59.1 |
| 4 | 59 | 59 | 59 | 58 | 58 | 58 | 59 | 59 | 58.6 |
| 5 | 58 | 58 | 59 | 58 | 58 | 58 | 58 | 58 | 58.1 |
| 6 | 58 | 58 | 58 | 58 | 58 | 58 | 58 | 58 | 58.0 |
| 7 | 58 | 58 | 57 | 58 | 58 | 58 | 58 | 58 | 57.9 |
| C | 58 | .. | .. | .. | 58 | .. | .. | .. | 57.9 |

Surface Hardness Frequency

| Section Size | 63 | 62 | 61 | 60 | 59 | 58 | 57 | 56 | 55 | |
|---|---|---|---|---|---|---|---|---|---|---|
| ½ inch C | | | 5 | 18 | 9 | 8 | .. | .. | .. | |
| ½ inch D | | | 11 | 25 | 4 | .. | .. | .. | .. | |
| Surface hardness of ½-inch rounds | | | | | | | | | | 60 |
| 1 inch C | | | 21 | 14 | 4 | 1 | .. | .. | .. | |
| 1 inch D | | | 6 | 30 | 4 | .. | .. | .. | .. | |
| Surface hardness of 1-inch rounds | | | | | | | | | | 60 |
| 2 inch 1A-3 | | 5 | 23 | 6 | 4 | 2 | .. | .. | .. | |
| 2 inch 1B-3 | | 10 | 25 | 5 | .. | .. | .. | .. | .. | |
| Surface hardness of 2-inch rounds | | | | | | | | | | 61 |
| 3 inch 1B-3 | | | 11 | 14 | 8 | 7 | .. | .. | .. | |
| 3 inch 1A-3 | | | 1 | 11 | 19 | 8 | 1 | .. | .. | |
| Surface hardness of 3-inch rounds | | | | | | | | | | 60 |

2-Inch Round

| | Specimen 1A-3 | | | | Specimen 1B-3 | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 59 | 60 | 60 | 60 | 61 | 60 | 60 | 59 | 59.9 |
| 2 | 59 | 60 | 60 | 59 | 59 | 59 | 60 | 59 | 59.4 |
| 3 | 59 | 60 | 61 | 59 | 58 | 59 | 59 | 58 | 59.1 |
| 4 | 59 | 59 | 60 | 59 | 58 | 59 | 59 | 58 | 58.9 |
| 5 | 58 | 58 | 60 | 59 | 58 | 59 | 58 | 58 | 58.5 |
| 6 | 57 | 57 | 59 | 58 | 58 | 59 | 58 | 57 | 57.9 |
| 7 | 56 | 56 | 59 | 58 | 57 | 58 | 57 | 56 | 57.1 |
| 8 | 56 | 55 | 58 | 58 | 56 | 56 | 56 | 55 | 56.3 |
| 9 | 55 | 55 | 58 | 56 | 55 | 55 | 54 | 54 | 55.3 |
| 10 | 54 | 55 | 56 | 55 | 54 | 54 | 53 | 54 | 54.4 |
| 11 | 55 | 54 | 54 | 55 | 53 | 53 | 53 | 54 | 53.9 |
| 12 | 53 | 53 | 53 | 54 | 53 | 52 | 52 | 53 | 52.9 |
| 13 | 52 | 53 | 53 | 52 | 52 | 52 | 52 | 53 | 52.4 |
| 14 | 51 | 51 | 53 | 52 | 52 | 51 | 52 | 52 | 51.8 |
| 15 | 51 | 51 | 51 | 52 | 51 | 50 | 51 | 52 | 51.1 |
| C | 50 | .. | .. | .. | 51 | .. | .. | .. | 51.0 |

The points on the transverse hardness distribution curves, except the center and surface values, are the average of 8 values, obtained on 1-inch sections taken from two different quenched specimens. The points on the tempering curves are average values obtained on one specimen only.

**Table VII (Continued)**

| | 3-Inch Round | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | Specimen 1A-3 | | | | Specimen 1B-3 | | | | |
| | 1 | 2 | 3 | 4 | 1 | 2 | 3 | 4 | Average |
| 1 | 59 | 59 | 60 | 59 | 59 | 59 | 60 | 59 | 59.3 |
| 2 | 60 | 58 | 60 | 59 | 58 | 57 | 59 | 58 | 58.6 |
| 3 | 58 | 59 | 59 | 58 | 58 | 56 | 59 | 56 | 57.9 |
| 4 | 58 | 57 | 59 | 57 | 57 | 55 | 58 | 55 | 57.0 |
| 5 | 56 | 57 | 57 | 56 | 56 | 56 | 57 | 55 | 56.3 |
| 6 | 54 | 55 | 57 | 56 | 54 | 54 | 56 | 53 | 54.9 |
| 7 | 52 | 52 | 57 | 55 | 53 | 53 | 55 | 53 | 53.8 |
| 8 | 50 | 51 | 55 | 52 | 51 | 52 | 53 | 53 | 52.1 |
| 9 | 49 | 48 | 52 | 51 | 48 | 49 | 52 | 51 | 50.0 |
| 10 | 47 | 48 | 48 | 48 | 46 | 48 | 48 | 50 | 47.9 |
| 11 | 45 | 47 | 48 | 46 | 45 | 45 | 46 | 46 | 46.0 |
| 12 | 44 | 45 | 45 | 43 | 43 | 43 | 44 | 44 | 43.9 |
| 13 | 44 | 44 | 46 | 43 | 43 | 43 | 43 | 43 | 43.6 |
| 14 | 43 | 44 | 43 | 42 | 42 | 43 | 42 | 41 | 42.5 |
| 15 | 42 | 43 | 42 | 43 | 42 | 42 | 42 | 41 | 42.1 |
| 16 | 42 | 43 | 42 | 41 | 41 | 42 | 42 | 41 | 41.8 |
| 17 | 42 | 43 | 42 | 42 | 40 | 41 | 40 | 41 | 41.4 |
| 18 | 44 | 41 | 41 | 41 | 42 | 42 | 39 | 40 | 41.3 |
| 19 | 42 | 40 | 41 | 41 | 40 | 41 | 41 | 40 | 40.8 |
| 20 | 42 | 42 | 40 | 41 | 41 | 40 | 39 | 41 | 40.8 |
| 21 | 42 | 41 | 41 | 42 | 40 | 41 | 40 | 41 | 41.0 |
| 22 | 41 | 43 | 41 | 41 | 41 | 40 | 40 | 42 | 41.1 |
| 23 | 40 | 41 | 41 | 42 | 39 | 41 | 42 | 41 | 40.9 |
| C | 41 | .. | .. | .. | 40 | .. | .. | .. | 40.8 |

Note—All columns of penetration readings start with the hardness $\frac{1}{8}$ inch from surface and proceed downward to the hardness at the center. Readings taken every $\frac{1}{8}$ of an inch along a radius. Average center hardness is the average of center and values on a circle $\frac{1}{8}$ inch away from center.

## Discussion of Results

*Hardenability of Water-Quenched Steels*—Maximum surface hardness was obtained on all section sizes for each steel, showing that the critical cooling rate had been reached or exceeded on water quenching the steels. The maximum hardness obtained for the 0.30 per cent carbon steels was very close to 55 Rockwell "C", while for the 0.45 per cent carbon steels the maximum hardness obtained was 61 to 62 Rockwell "C". These values are in close agreement with the curve showing the dependence of maximum hardness on the carbon content that appeared in the paper by Burns, Moore, and Archer (1).

The plain carbon steels of S.A.E. 1045 grade have essentially the same chemical composition but differ in that one is a fine-grained and the other is a coarse-grained steel. The former is definitely more shallow hardening than the latter.

It was interesting to find that the contour of the curves as well as the slight increase in hardness near the center for the fine-grained steel was almost identical with curves obtained by Burns, Moore and Archer (1) for a steel of the same grade.

**Table VIII**

**Brinell and Rockwell C Hardness Values of Oil-Quenched and Tempered S.A.E. 6145 Steel**

| Section Size | Rockwell "C" Hardness Values: Center Values: Center | 1 | 2 | 3 | 4 | Avg. | Values for ½ Distance Between Center and Edge: 1 | 2 | 3 | 4 | Avg. | Mode of Surface Frequency | Brinell Hardness Values*: Center Values | Values for ½ Distance Between Center and Edge: 1 | 2 | 3 | 4 | Avg. | Surface Values: 1 | 2 | 3 | 4 | Avg. |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | Tempered at 400 Degrees Fahr. | | | | | | | | | | | | | | | | | | | | | | |
| 1 inch C | 53 | 54 | 54 | 54 | 54 | 53.8 | 55 | 56 | 56 | 55 | 55.5 | 56 | 578 | 578 | 592 | 592 | 601 | 591 | 601 | 601 | 611 | 601 | 604 |
| 1 inch D | 53 | 53 | 53 | 54 | 53 | 53.2 | 54 | 53 | 53 | 55 | 53.8 | 55 | 578 | 578 | 578 | 578 | 578 | 578 | 601 | 601 | 601 | 601 | 601 |
| 2 inch 1A-2 | 49 | 50 | 50 | 49 | 51 | 49.8 | 55 | 54 | 55 | 54 | 54.5 | 55 | 514 | 601 | 578 | 601 | 587 | 592 | 601 | 601 | 601 | 601 | 601 |
| 2 inch 1B-2 | 49 | 50 | 50 | 50 | 49 | 49.6 | 54 | 54 | 53 | 54 | 53.8 | 55 | 514 | 592 | 578 | 601 | 578 | 587 | 601 | 601 | 601 | 601 | 601 |
| 3 inch 1A-2 | 40 | 42 | 43 | 42 | 42 | 41.8 | 46 | 46 | 45 | 46 | 45.8 | 56 | 415 | 471 | 461 | 461 | 451 | 461 | 601 | 601 | 601 | 601 | 601 |
| 3 inch 1B-2 | 39 | 41 | 42 | 42 | 42 | 41.2 | 43 | 42 | 43 | 43 | 42.8 | 55 | 415 | 451 | 461 | 451 | 461 | 456 | 611 | 601 | 601 | 601 | 601 |
| | Tempered at 800 Degrees Fahr. | | | | | | | | | | | | | | | | | | | | | | |
| 1 inch C | 41 | 43 | 42 | 44 | 43 | 42.6 | 44 | 44 | 44 | 44 | 44.0 | 45 | 444 | 471 | 461 | 461 | 461 | 464 | 477 | 477 | 471 | 471 | 474 |
| 1 inch D | 41 | 43 | 42 | 44 | 43 | 42.6 | 43 | 44 | 44 | 44 | 43.8 | 45 | 451 | 461 | 461 | 471 | 461 | 464 | 477 | 477 | 477 | 477 | 477 |
| 2 inch 1A-3 | 41 | 42 | 43 | 43 | 43 | 42.4 | 43 | 44 | 44 | 45 | 44.0 | 45 | 444 | 471 | 471 | 477 | 477 | 474 | 485 | 477 | 485 | 477 | 481 |
| 2 inch 1B-3 | 41 | 43 | 42 | 42 | 42 | 42.0 | 43 | 45 | 43 | 44 | 43.8 | 45 | 444 | 477 | 477 | 477 | 461 | 473 | 485 | 477 | 485 | 477 | 481 |
| 3 inch 1A-3 | 37 | 37 | 38 | 38 | 39 | 37.8 | 39 | 40 | 40 | 40 | 39.8 | 45 | 388 | 415 | 415 | 415 | 415 | 415 | 477 | 477 | 477 | 477 | 477 |
| 3 inch 1B-3 | 37 | 37 | 37 | 38 | 38 | 37.4 | 39 | 39 | 40 | 40 | 39.5 | 44 | 388 | 415 | 406 | 406 | 415 | 411 | 477 | 477 | 485 | 477 | 479 |
| | Tempered at 1000 Degrees Fahr. | | | | | | | | | | | | | | | | | | | | | | |
| 1 inch C | 39 | 39 | 39 | 39 | 39 | 39.0 | 40 | 39 | 40 | 39 | 39.5 | 40 | 388 | 388 | 388 | 388 | 388 | 388 | 415 | 401 | 401 | 401 | 405 |
| 1 inch D | 40 | 40 | 39 | 39 | 40 | 39.6 | 40 | 40 | 40 | 40 | 40.0 | 40 | 393 | 393 | 393 | 393 | 401 | 395 | 409 | 409 | 415 | 409 | 411 |
| 2 inch 1A-4 | 38 | 38 | 38 | 38 | 38 | 38.0 | 40 | 40 | 40 | 39 | 39.8 | 40 | 383 | 393 | 409 | 393 | 393 | 397 | 415 | 415 | 415 | 415 | 415 |
| 2 inch 1B-4 | 38 | 39 | 38 | 38 | 39 | 38.4 | 41 | 40 | 40 | 40 | 40.3 | 41 | 383 | 393 | 415 | 401 | 406 | 404 | 415 | 415 | 415 | 415 | 415 |
| 3 inch 1A-4 | 32 | 34 | 34 | 34 | 35 | 33.8 | 36 | 35 | 34 | 35 | 35.0 | 39 | 341 | 363 | 363 | 363 | 363 | 363 | 415 | 415 | 415 | 415 | 415 |
| 3 inch 1B-4 | 35 | 34 | 34 | 34 | 34 | 34.2 | 37 | 36 | 35 | 36 | 36.0 | 40 | 352 | 363 | 363 | 363 | 363 | 363 | 415 | 415 | 409 | 415 | 413 |

*Tungsten carbide ball.

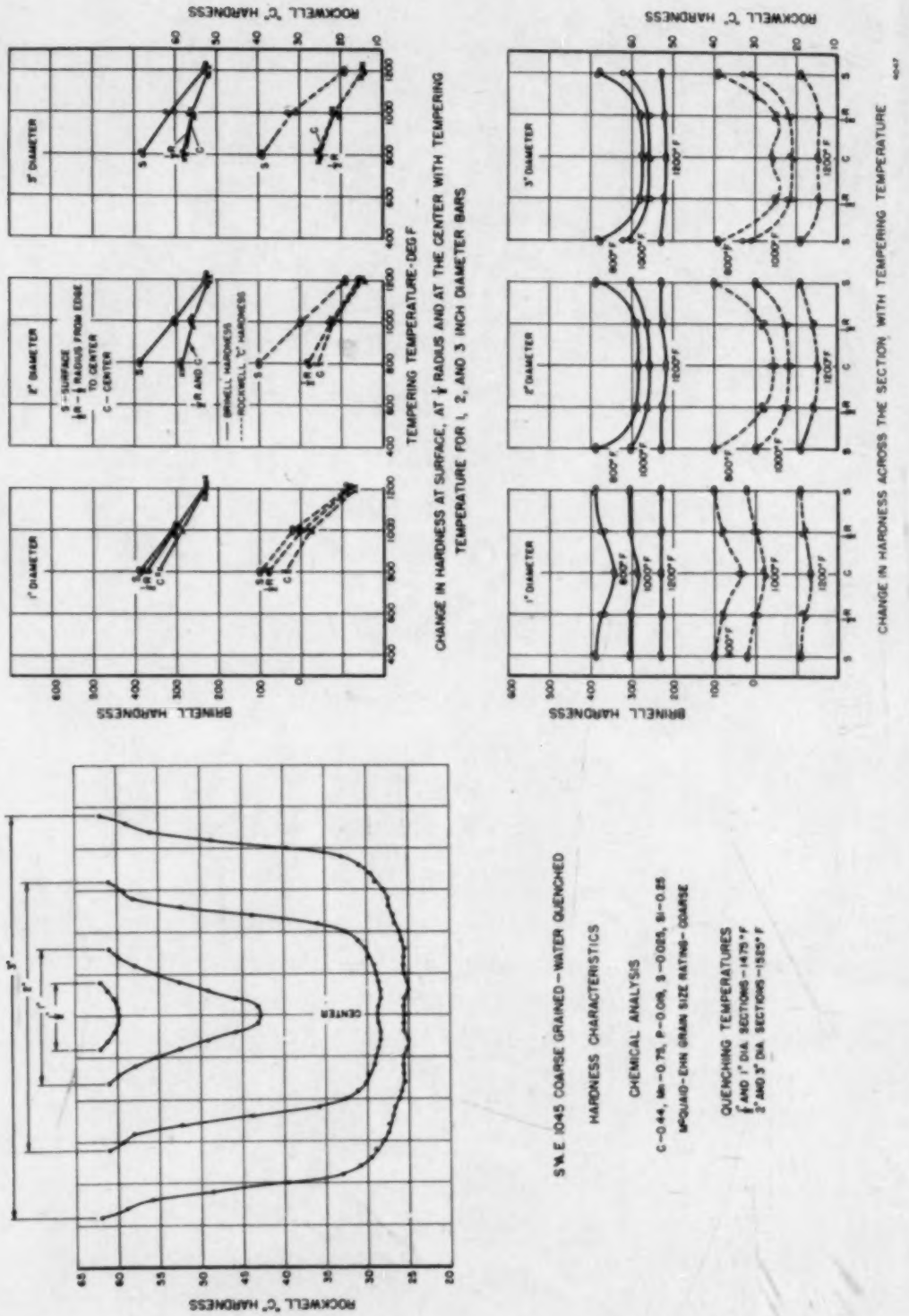
ROCKWELL 'C' HARDNESS
BRINELL HARDNESS
TEMPERING TEMPERATURE-DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
S-SURFACE ½R-½ RADIUS FROM EDGE C-CENTER
BRINELL HARDNESS
ROCKWELL 'C' HARDNESS
CENTER
S.A.E. 1045 COARSE GRAINED-WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.44, Mn-0.75, P-0.018, S-0.025, Si-0.25
SQUARE-EHN GRAIN SIZE RATING-COARSE
QUENCHING TEMPERATURES
½" AND 1" DIA. SECTIONS-1475°F
2" AND 3" DIA. SECTIONS-1525°F

Fig. 5—Depth Hardness Characteristics of Water-Quenched S.A.E. 1045 Coarse-Grained Steel.

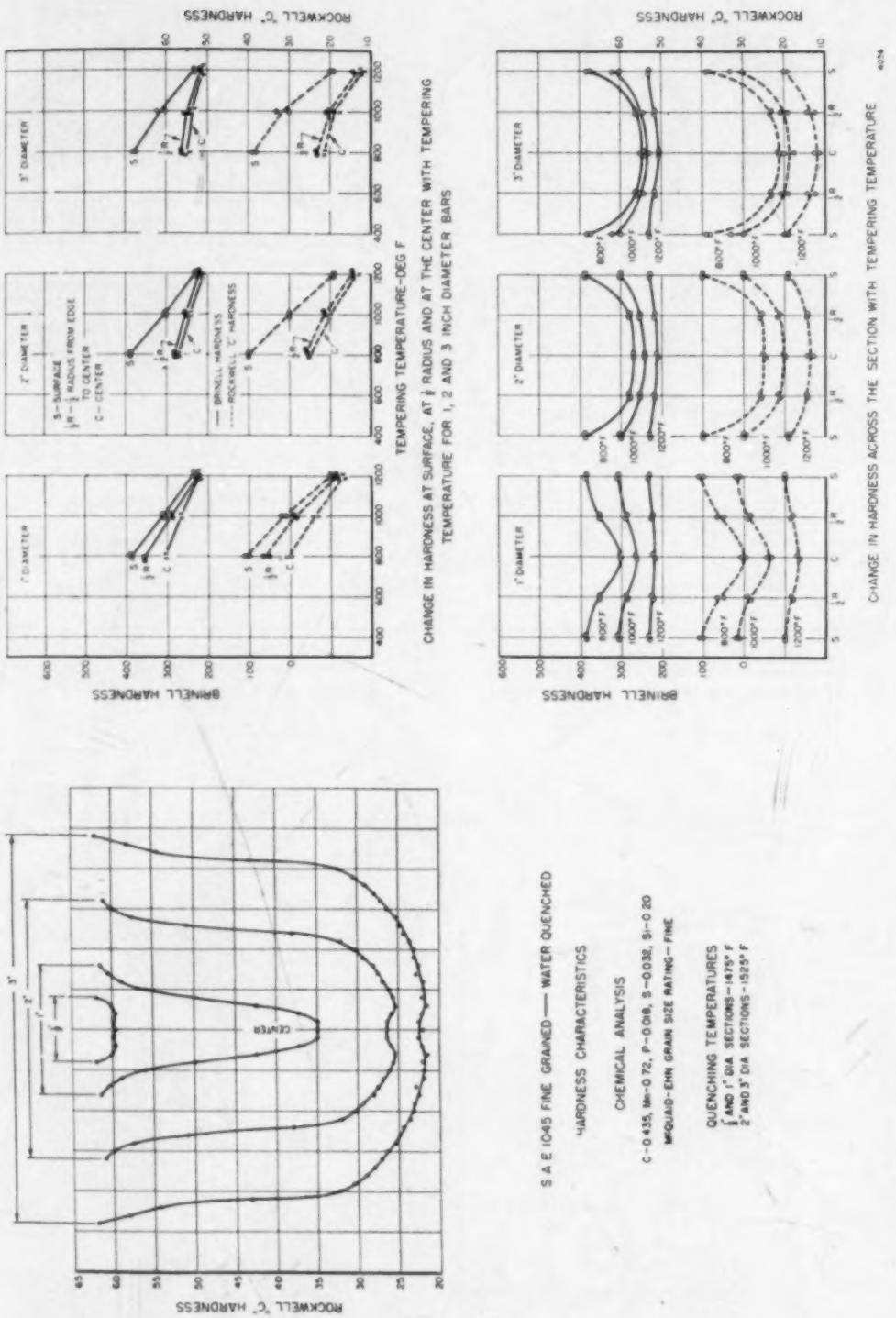
ROCKWELL "C" HARDNESS
3" DIAMETER
2" DIAMETER
1" DIAMETER
S—SURFACE
½R—½ RADIUS FROM EDGE TO CENTER
C—CENTER
BRINELL HARDNESS
ROCKWELL "C" HARDNESS
TEMPERING TEMPERATURE—DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2 AND 3 INCH DIAMETER BARS
BRINELL HARDNESS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
800°F
1000°F
1200°F
CENTER
S A E 1045 FINE GRAINED — WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.435, Mn-0.72, P-0.018, S-0.032, Si-0.20
McQUAID-EHN GRAIN SIZE RATING—FINE
QUENCHING TEMPERATURES
½" AND 1" DIA SECTIONS—1475° F
2" AND 3" DIA SECTIONS—1525° F

Fig. 6—Depth Hardness Characteristics of Water-Quenched S.A.E. 1045 Fine-Grained Steel.

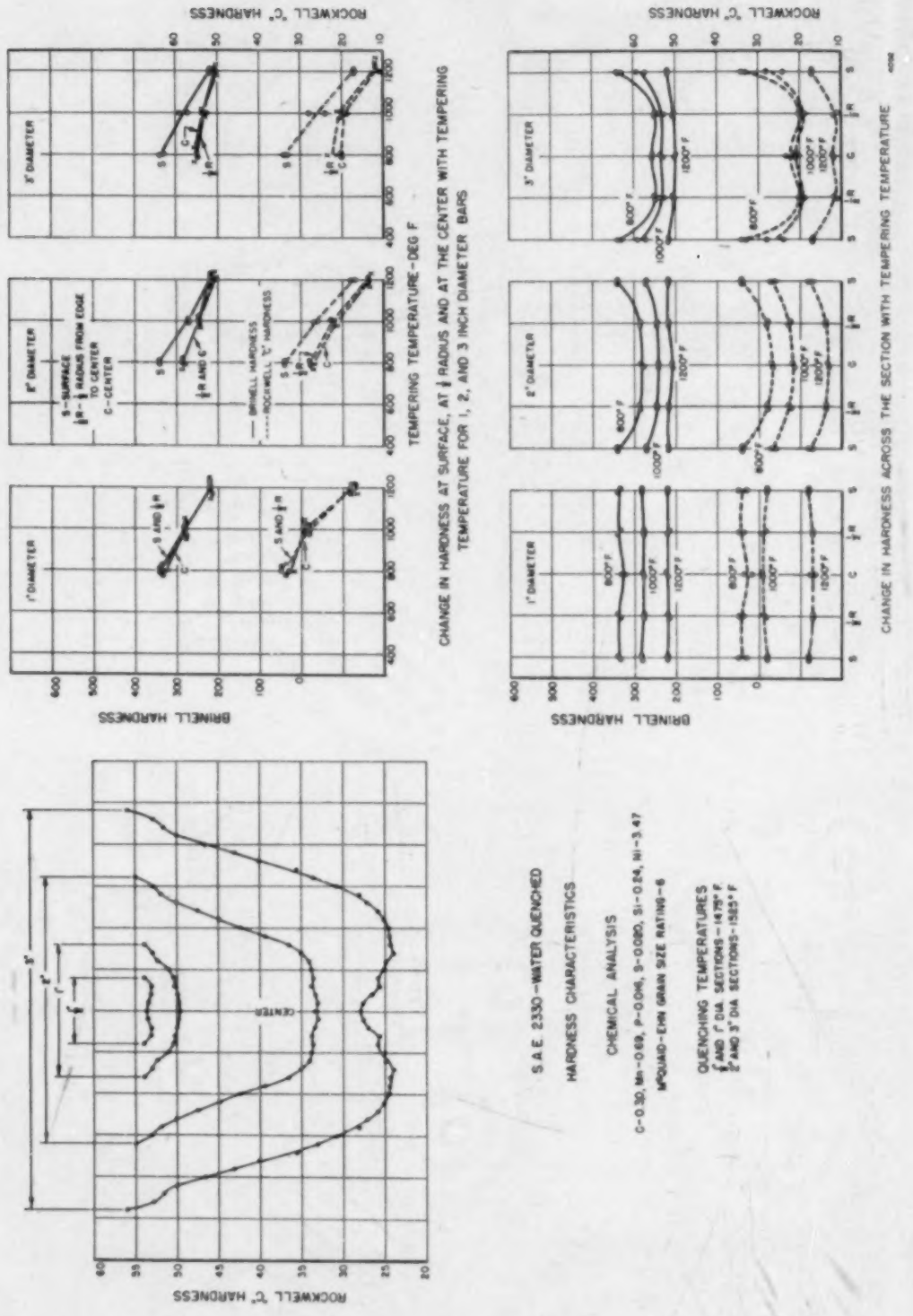
ROCKWELL "C" HARDNESS
BRINELL HARDNESS
TEMPERING TEMPERATURE - DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
S A E 2330 — WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.30, Mn-0.68, P-0.014, S-0.020, Si-0.24, Ni-3.47
McQUAID-EHN GRAIN SIZE RATING-6
QUENCHING TEMPERATURES
½" AND 1" DIA SECTIONS - 1475° F
2" AND 3" DIA SECTIONS - 1525° F
CENTER

Fig. 7—Depth Hardness Characteristics of Water-Quenched S.A.E. 2330 Steel.

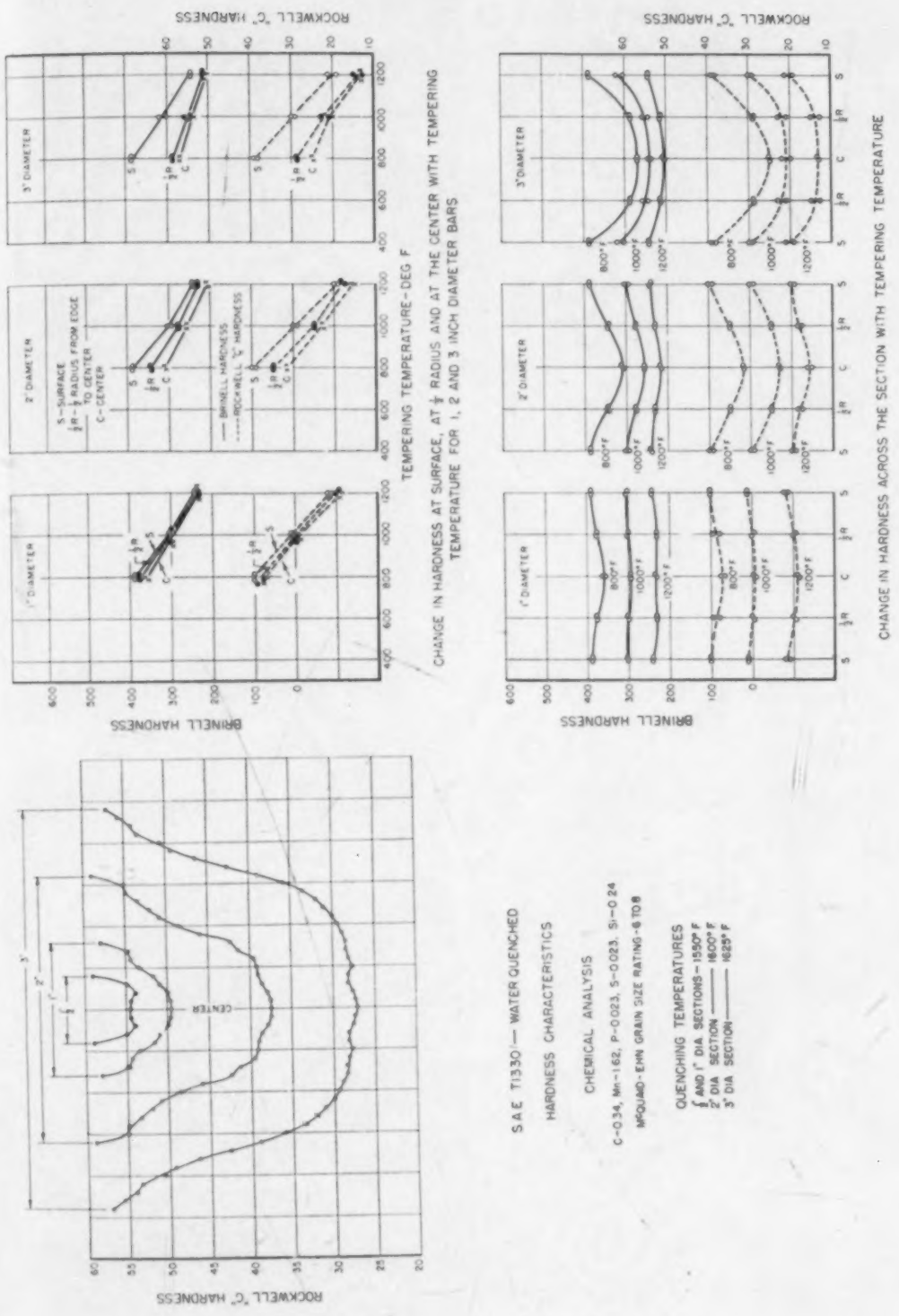
ROCKWELL "C" HARDNESS
3" DIAMETER
2" DIAMETER
1" DIAMETER
S—SURFACE
½R—½ RADIUS FROM EDGE TO CENTER
C—CENTER
BRINELL HARDNESS
ROCKWELL "C" HARDNESS
TEMPERING TEMPERATURE—DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2 AND 3 INCH DIAMETER BARS
BRINELL HARDNESS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
800°F
1000°F
1200°F
CENTER
S.A.E. T1330 — WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.34, Mn-1.62, P-0.023, S-0.023, Si-0.24
McQUAID-EHN GRAIN SIZE RATING-6 TO 8
QUENCHING TEMPERATURES
½" AND 1" DIA SECTIONS—1550° F
2" DIA SECTION ——— 1600° F
3" DIA SECTION ——— 1625° F

Fig. 8—Depth Hardness Characteristics of Water-Quenched S.A.E. T-1330 Steel.

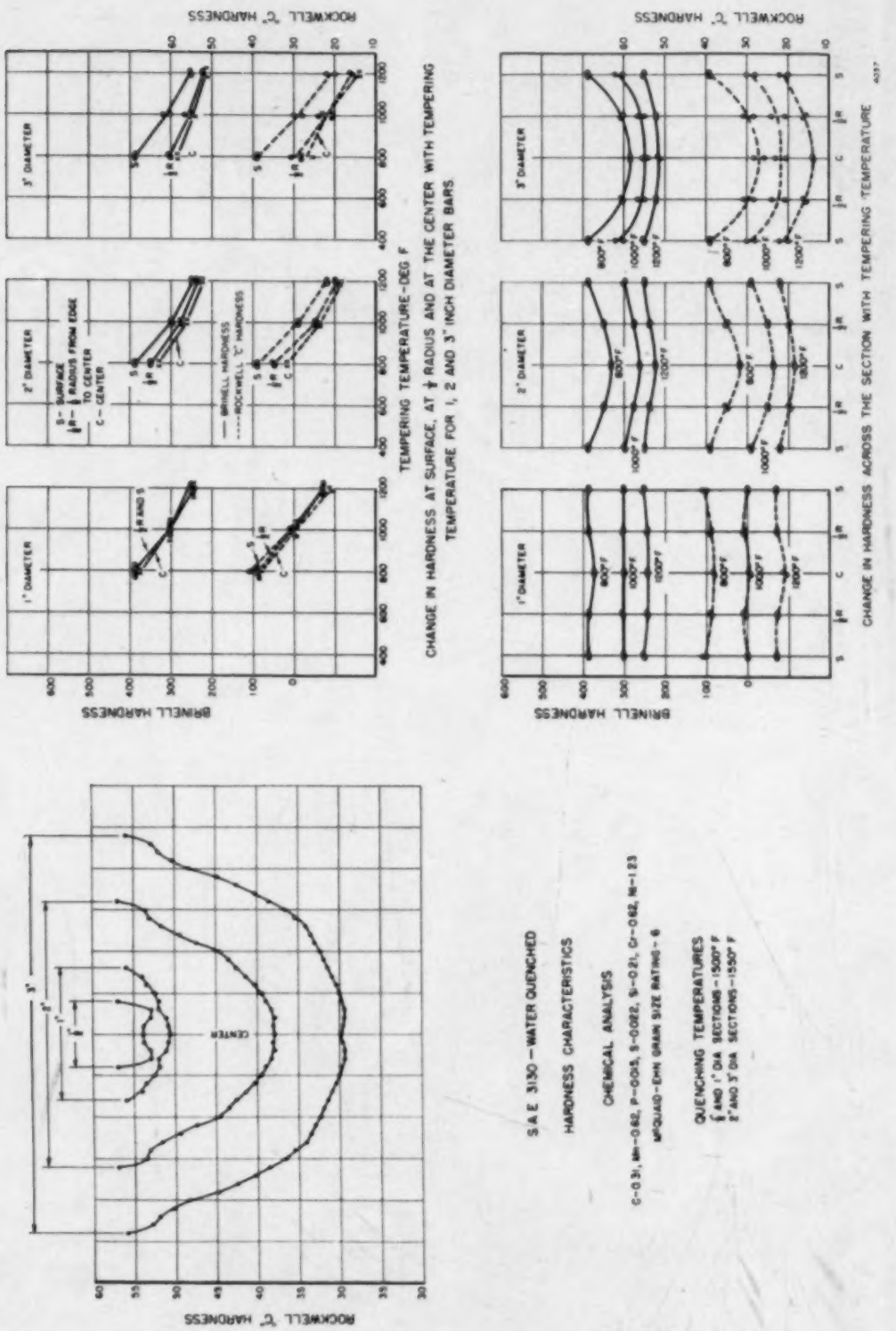
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2 AND 3" INCH DIAMETER BARS
TEMPERING TEMPERATURE-DEG F
ROCKWELL 'C' HARDNESS
BRINELL HARDNESS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
S.A.E. 3130 — WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSES
QUENCHING TEMPERATURES
CENTER

Fig. 9—Depth Hardness Characteristics of Water-Quenched S.A.E. 3130 Steel.

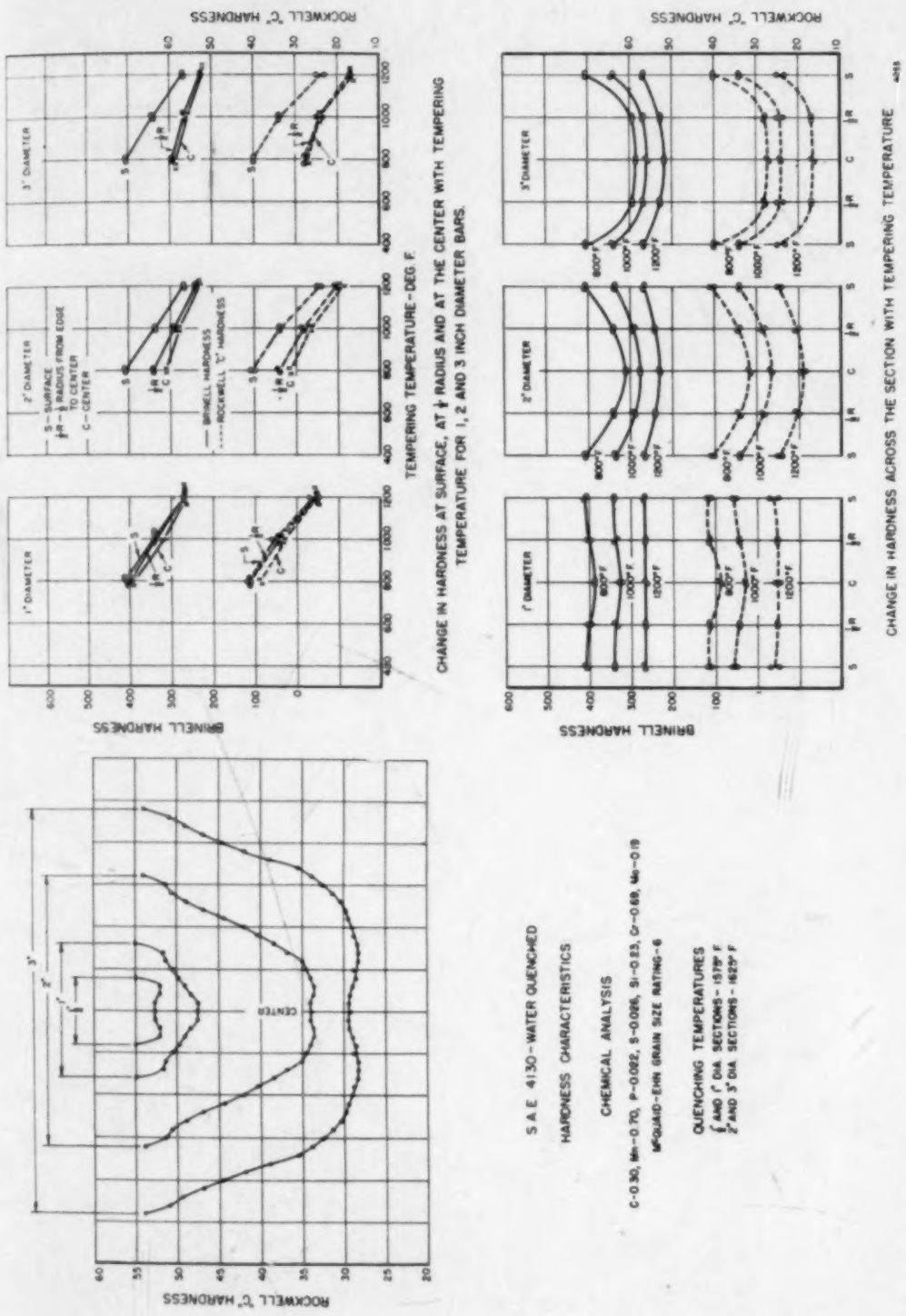
ROCKWELL "C" HARDNESS
3" DIAMETER
2" DIAMETER
1" DIAMETER
S – SURFACE
½R – ½ RADIUS FROM EDGE TO CENTER
C – CENTER
BRINELL HARDNESS
ROCKWELL "C" HARDNESS
TEMPERING TEMPERATURE – DEG F.
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2 AND 3 INCH DIAMETER BARS.
BRINELL HARDNESS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
CENTER
ROCKWELL "C" HARDNESS
S A E 4130 - WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.30, Mn-0.70, P-0.022, S-0.026, Si-0.23, Cr-0.68, Mo-0.19
McQUAID-EHN GRAIN SIZE RATING-6
QUENCHING TEMPERATURES
1" AND 1" DIA. SECTIONS - 1575° F
2" AND 3" DIA. SECTIONS - 1625° F

Fig. 10—Depth Hardness Characteristics of Water-Quenched S.A.E. 4130 Steel.

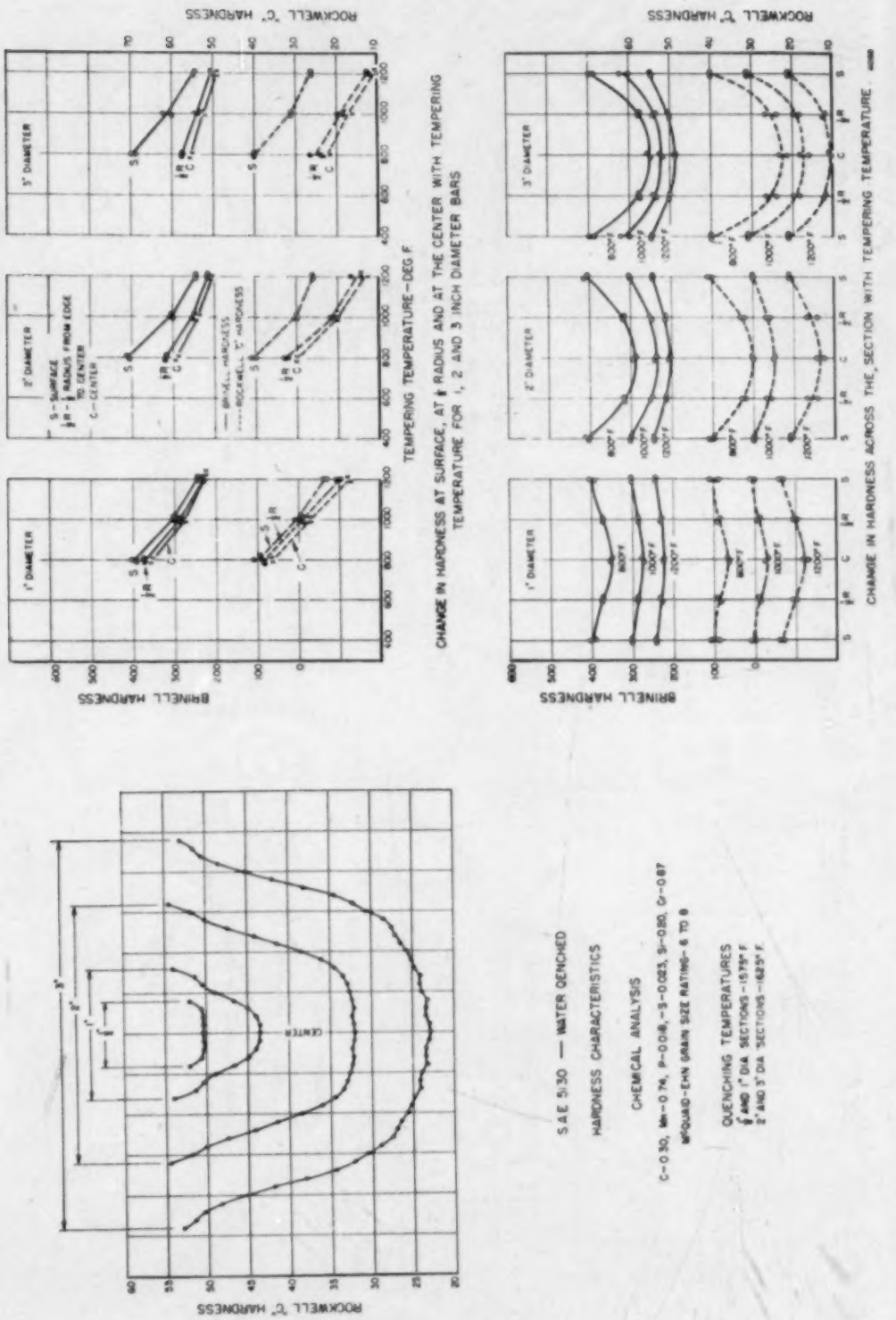
ROCKWELL 'C' HARDNESS
BRINELL HARDNESS
TEMPERING TEMPERATURE—DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2 AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
3" DIAMETER
2" DIAMETER
1" DIAMETER
S—SURFACE
½R—½ RADIUS FROM EDGE TO CENTER
C—CENTER
BRINELL HARDNESS
ROCKWELL 'C' HARDNESS
SAE 5130 — WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
QUENCHING TEMPERATURES

Fig. 11—Depth Hardness Characteristics of Water-Quenched S.A.E. 5130 Steel.

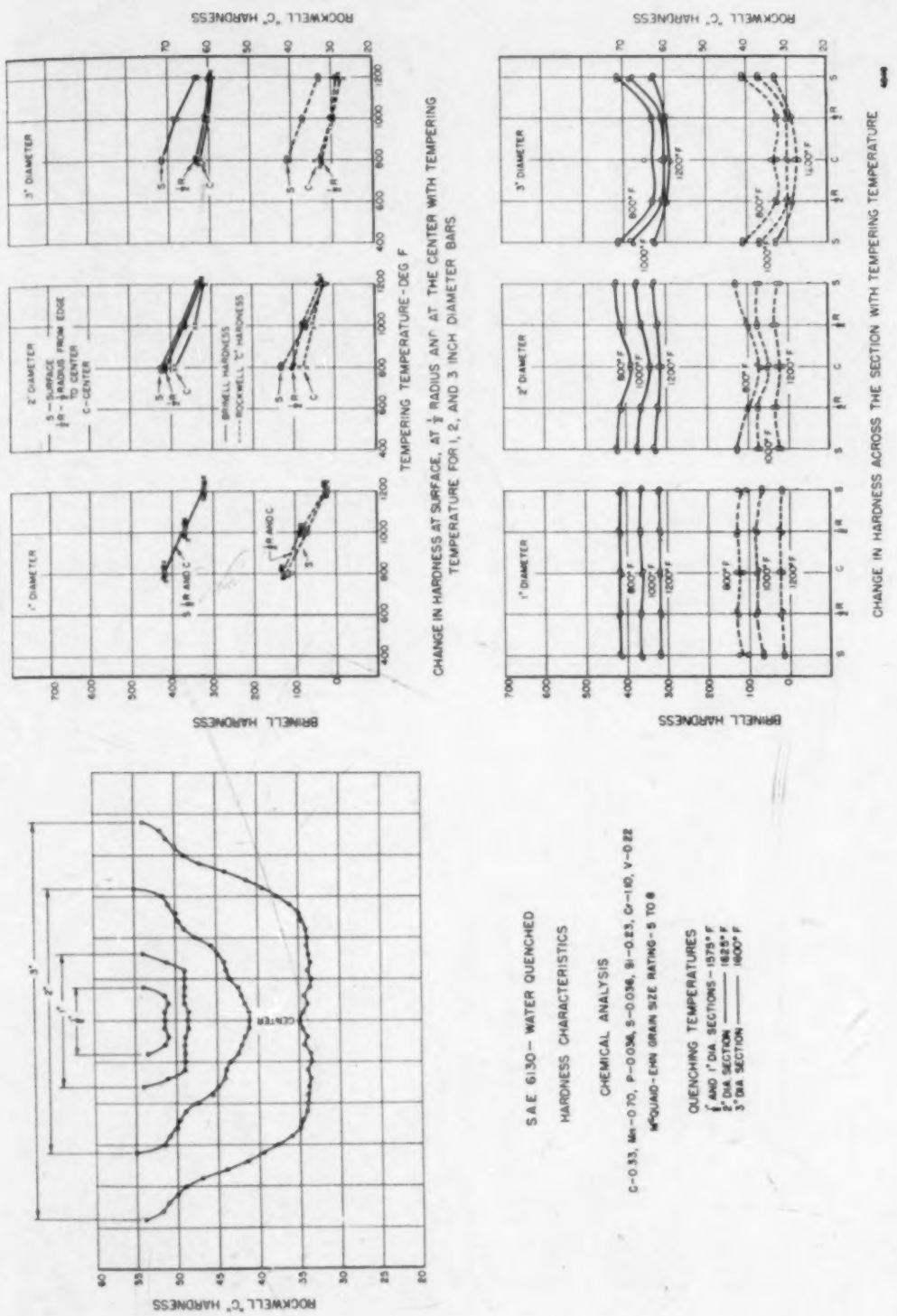
ROCKWELL "C" HARDNESS
3" DIAMETER
2" DIAMETER
1" DIAMETER
S — SURFACE
½R — ½ RADIUS FROM EDGE
C — CENTER
BRINELL HARDNESS
ROCKWELL "C" HARDNESS
BRINELL HARDNESS
TEMPERING TEMPERATURE - DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
SAE 6130 — WATER QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.33, Mn-0.70, P-0.036, S-0.036, Si-0.23, Cr-1.10, V-0.22
SQUARE-END GRAIN SIZE RATING-5 TO 8
QUENCHING TEMPERATURES
½" AND 1" DIA SECTIONS — 1575° F
2" DIA SECTION —— 1625° F
3" DIA SECTION —— 1600° F
CENTER
ROCKWELL "C" HARDNESS

Fig. 12—Depth Hardness Characteristics of Water-Quenched S.A.E. 6130 Steel.

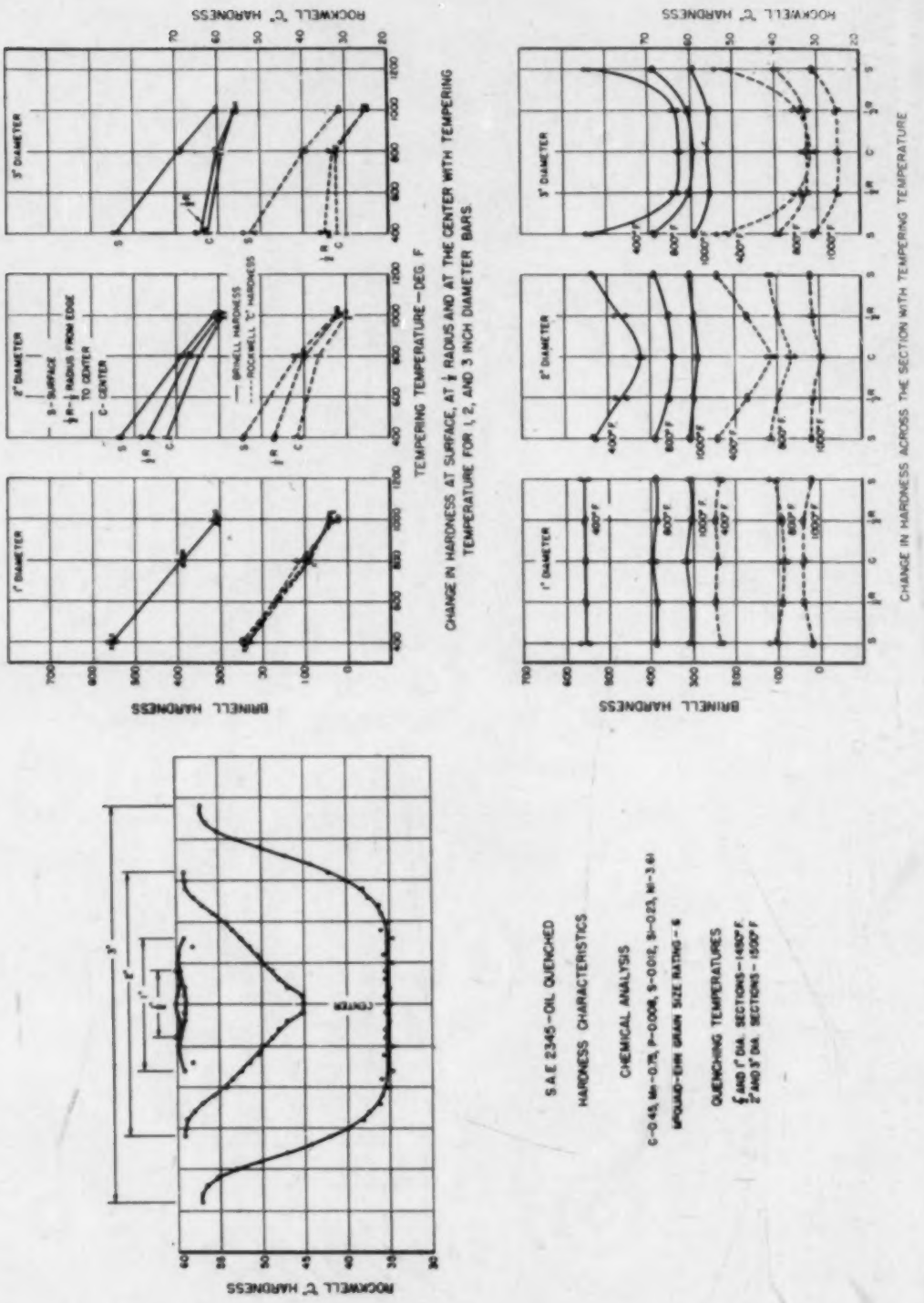
ROCKWELL C HARDNESS
BRINELL HARDNESS
TEMPERING TEMPERATURE—DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
SAE 2345—OIL QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
QUENCHING TEMPERATURES
CENTER

Fig. 13—Depth Hardness Characteristics of Oil-Quenched S.A.E. 2345 Steel.

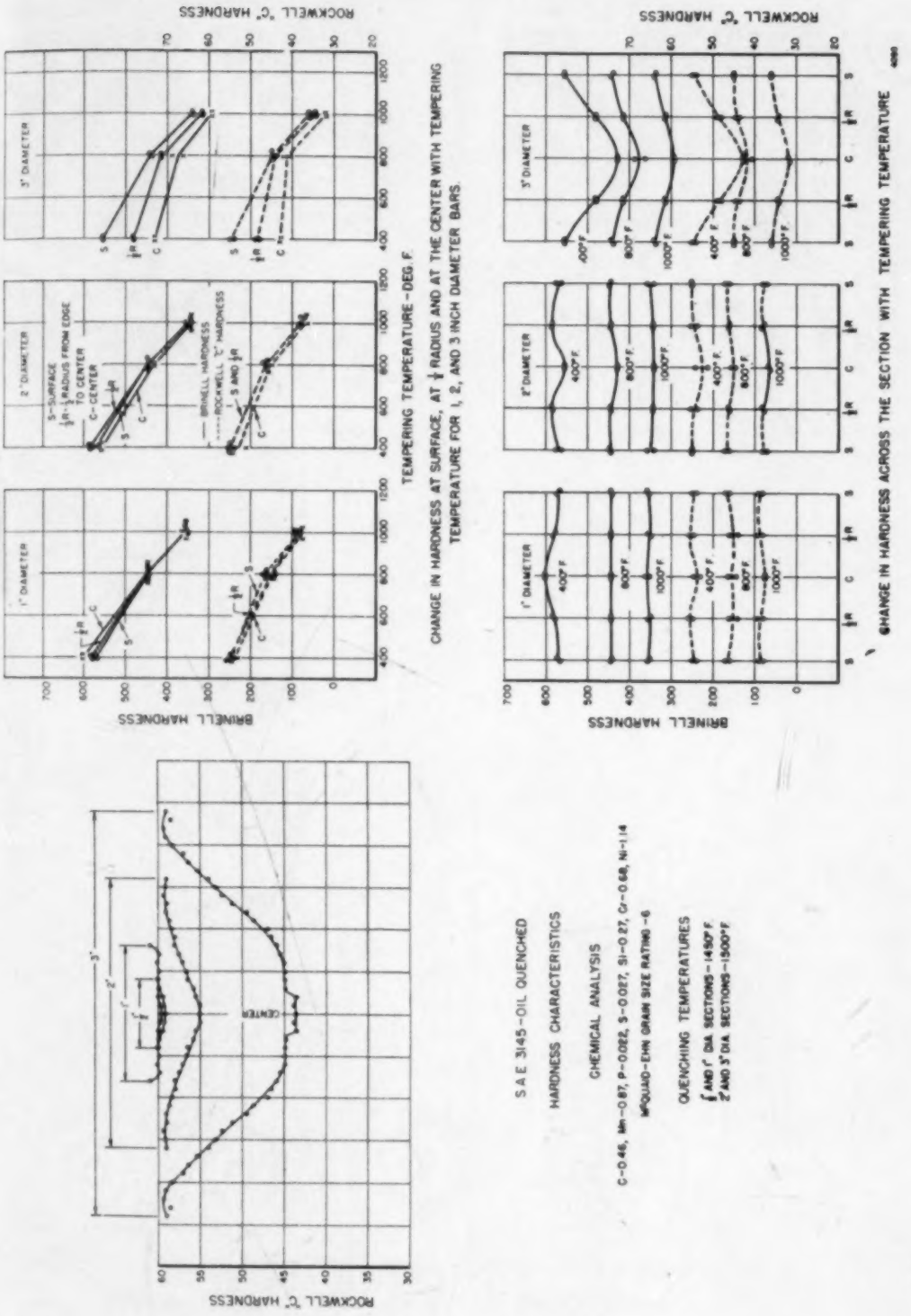


Fig. 14—Depth Hardness Characteristics of Oil-Quenched S.A.E. 3145 Steel.

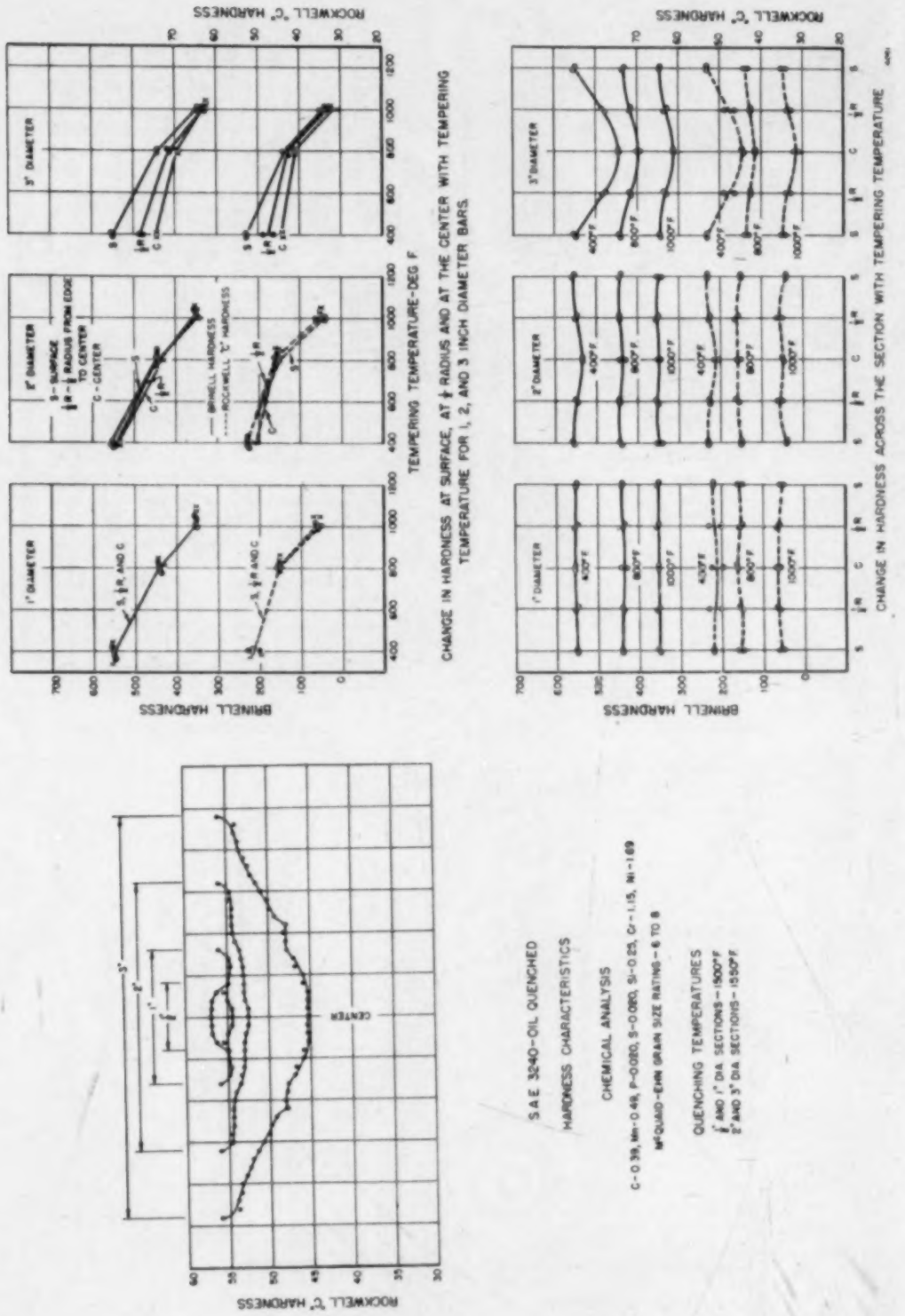


Fig. 15—Depth Hardness Characteristics of Oil-Quenched S.A.E. 3240 Steel.

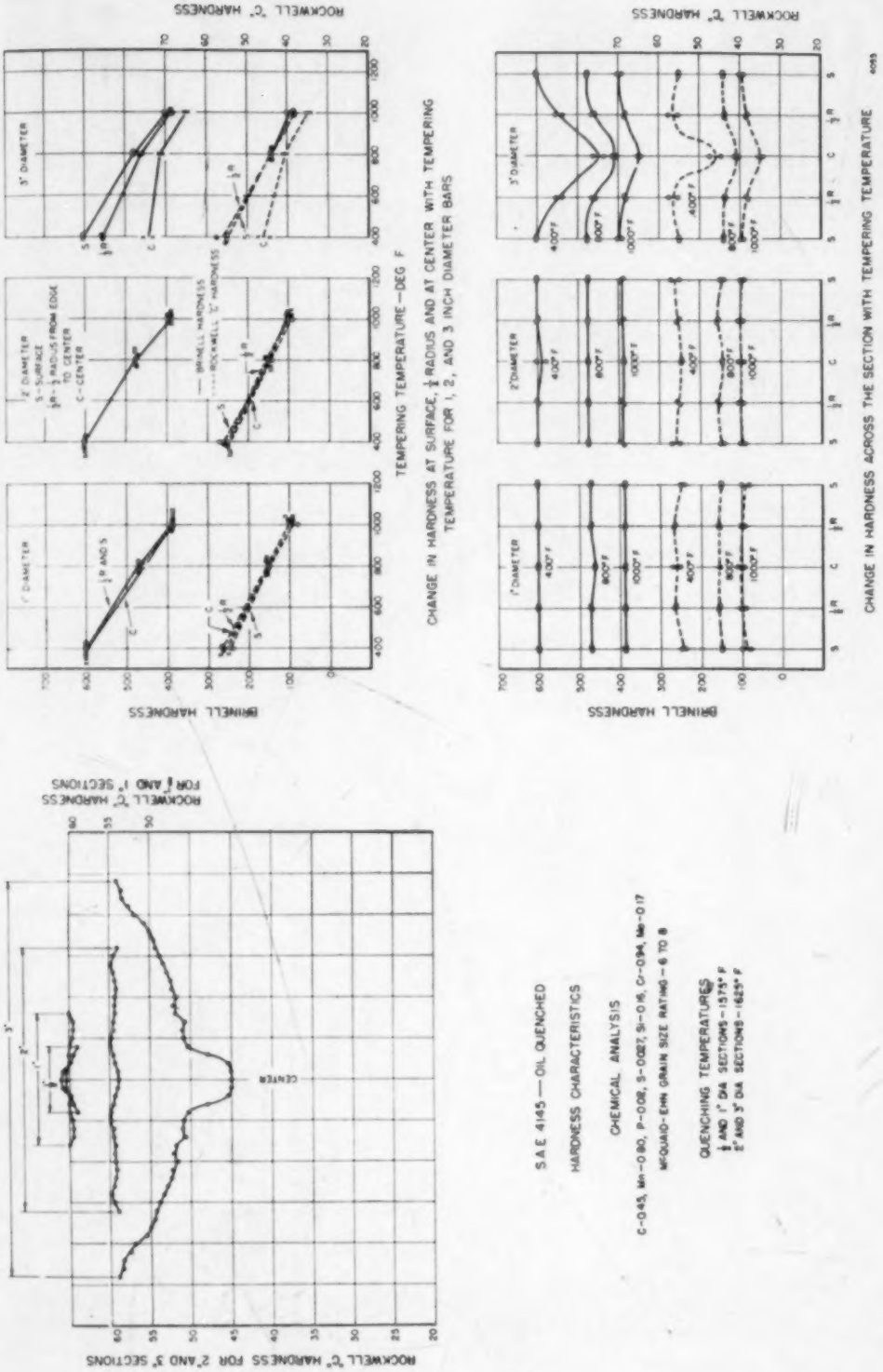
ROCKWELL 'C' HARDNESS
BRINELL HARDNESS
TEMPERING TEMPERATURE—DEG F
CHANGE IN HARDNESS AT SURFACE, ½ RADIUS AND AT CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
S.A.E. 4145 — OIL QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-.43, Mn-.80, P-.02, S-.027, Si-.15, Cr-.94, Ni-.17
MOLYBDENUM-GRAIN SIZE RATING-6 TO 8
QUENCHING TEMPERATURES
½ AND 1" DIA SECTIONS-1575° F
2" AND 3" DIA SECTIONS-1625° F
ROCKWELL 'C' HARDNESS FOR ½" AND 1" SECTIONS
ROCKWELL 'C' HARDNESS FOR 2" AND 3" SECTIONS
CENTER

Fig. 16—Depth Hardness Characteristics of Oil-Quenched S.A.E. 4145 Steel.

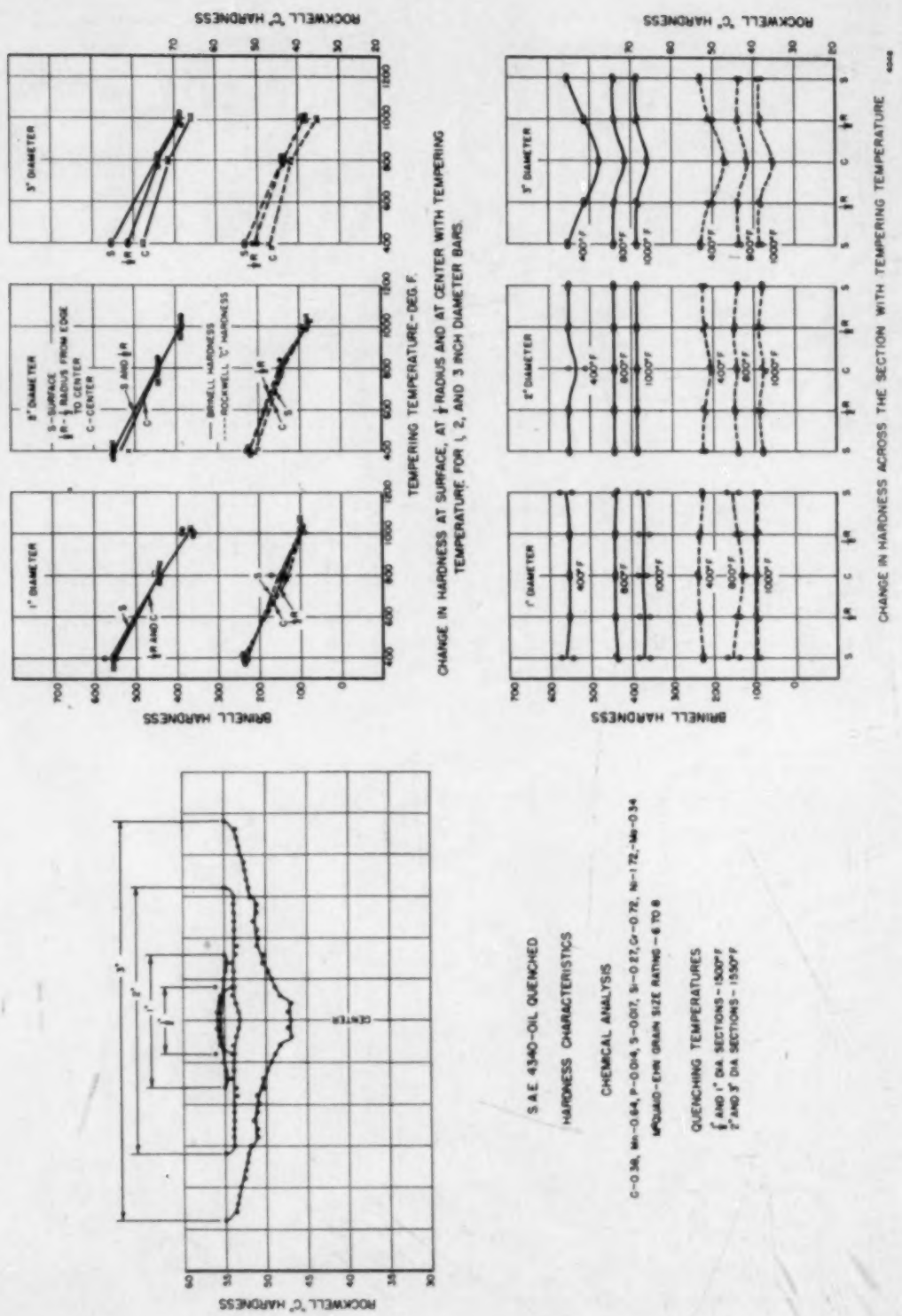
ROCKWELL 'C' HARDNESS
BRINELL HARDNESS
TEMPERING TEMPERATURE - DEG. F.
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
S.A.E. 4340-OIL QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
QUENCHING TEMPERATURES
CENTER

Fig. 17—Depth Hardness Characteristics of Oil-Quenched S.A.E. 4340 Steel.

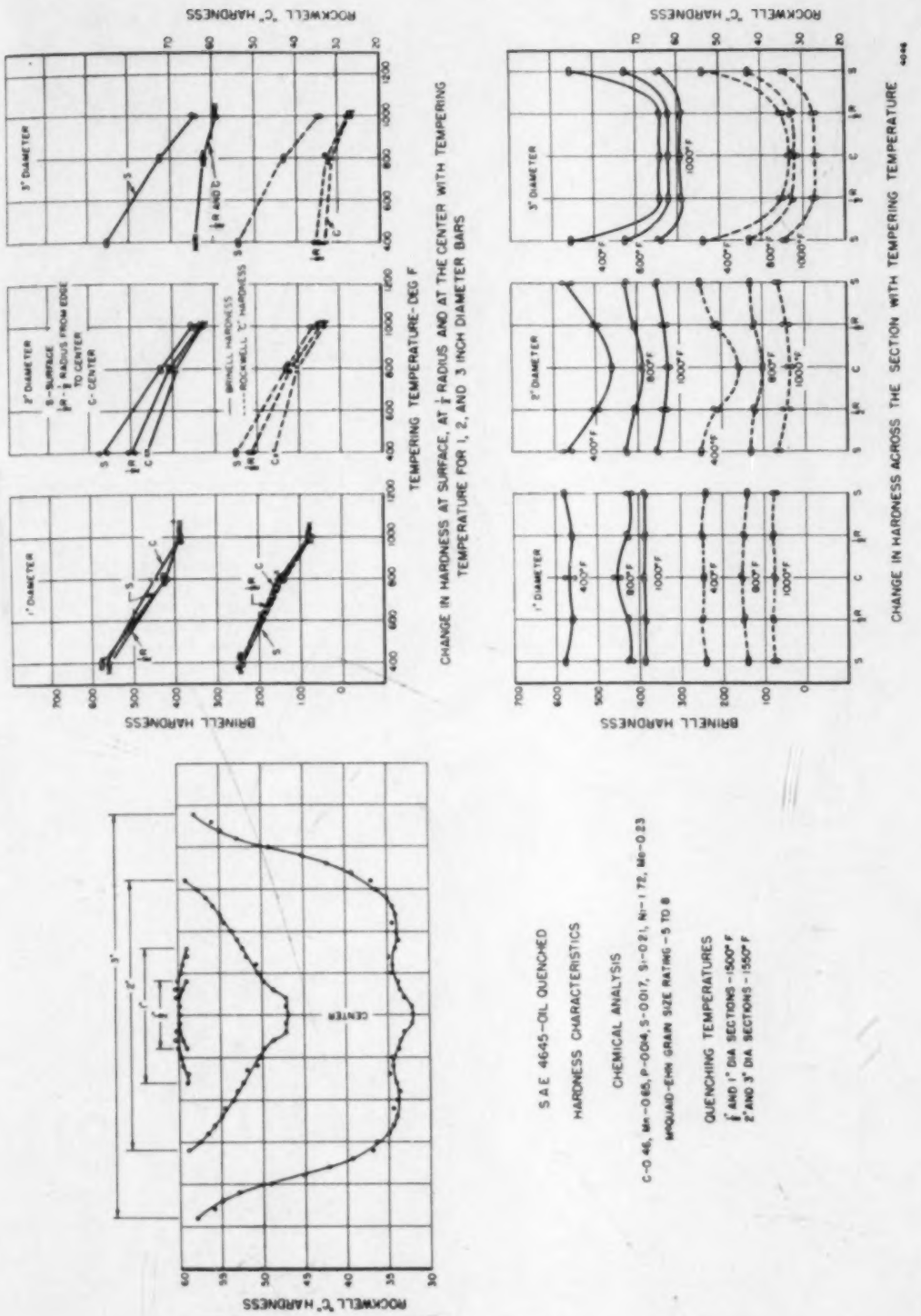
ROCKWELL "C" HARDNESS
3" DIAMETER
2" DIAMETER
1" DIAMETER
S - SURFACE
½R - ½ RADIUS FROM EDGE TO CENTER
C - CENTER
BRINELL HARDNESS
ROCKWELL "C" HARDNESS
TEMPERING TEMPERATURE - DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
400°F
800°F
1000°F
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
BRINELL HARDNESS
CENTER
SAE 4645-OIL QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.46, Mn-0.65, P-0.004, S-0.017, Si-0.21, Ni-1.72, Mo-0.23
NRQUAD-EHN GRAIN SIZE RATING - 5 TO 8
QUENCHING TEMPERATURES
½" AND 1" DIA SECTIONS - 1500° F
2" AND 3" DIA SECTIONS - 1550° F

Fig. 18—Depth Hardness Characteristics of Oil-Quenched S.A.E. 4645 Steel.

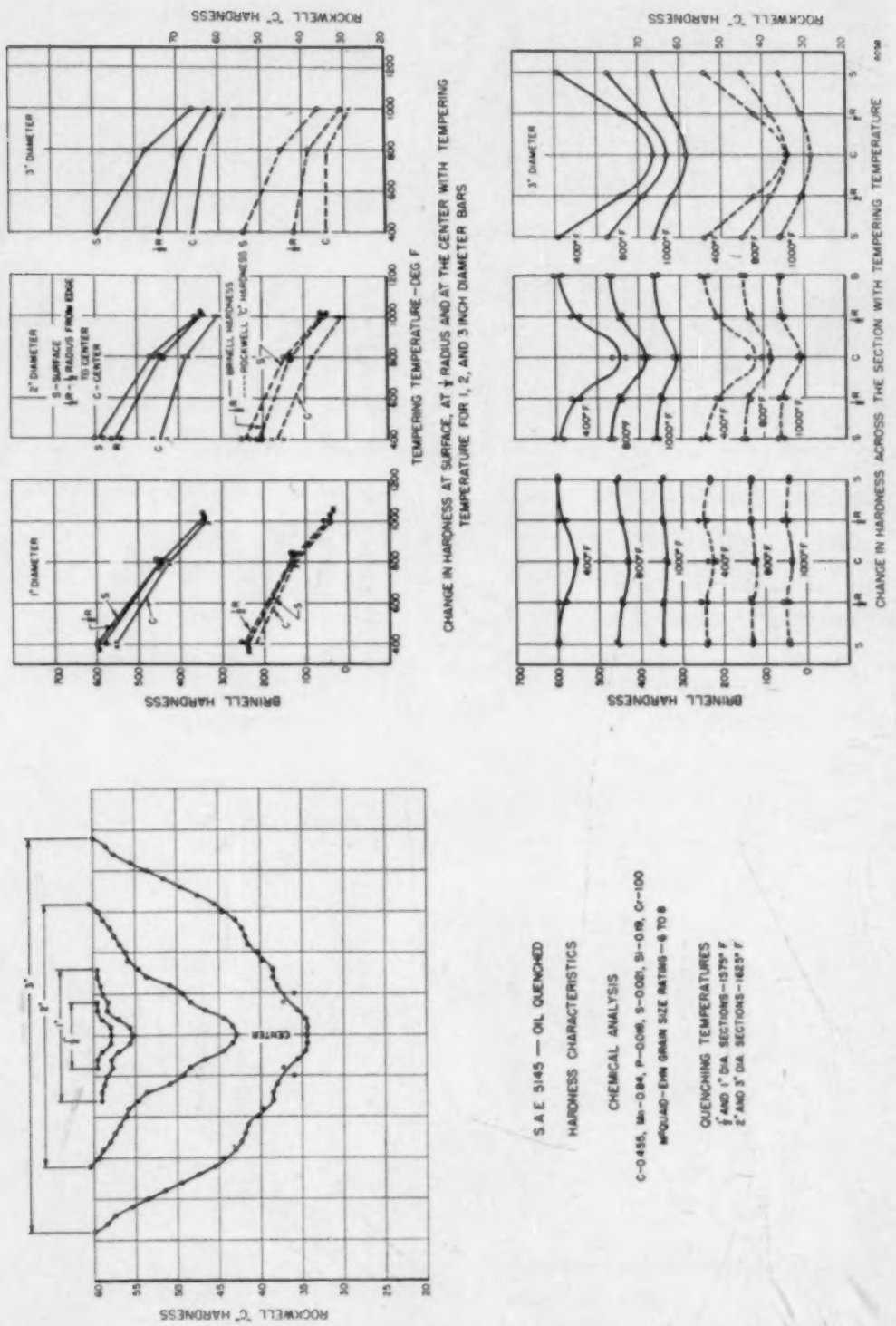
ROCKWELL "C" HARDNESS
3" DIAMETER
2" DIAMETER
S - SURFACE
½R - ½ RADIUS FROM EDGE TO CENTER
C - CENTER
BRINELL HARDNESS
ROCKWELL "C" HARDNESS
1" DIAMETER
BRINELL HARDNESS
TEMPERING TEMPERATURE - DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
400°F
800°F
1000°F
S.A.E. 5145 — OIL QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.45%, Mn-0.84, P-0.04, S-0.031, Si-0.19, Cr-1.00
MIQUAD-ERH GRAIN SIZE RATING-6 TO 8
QUENCHING TEMPERATURES
½" AND 1" DIA. SECTIONS-1575° F
2" AND 3" DIA. SECTIONS-1625° F
CENTER
ROCKWELL "C" HARDNESS

Fig. 19—Depth Hardness Characteristics of Oil-Quenched S.A.E. 5145 Steel.

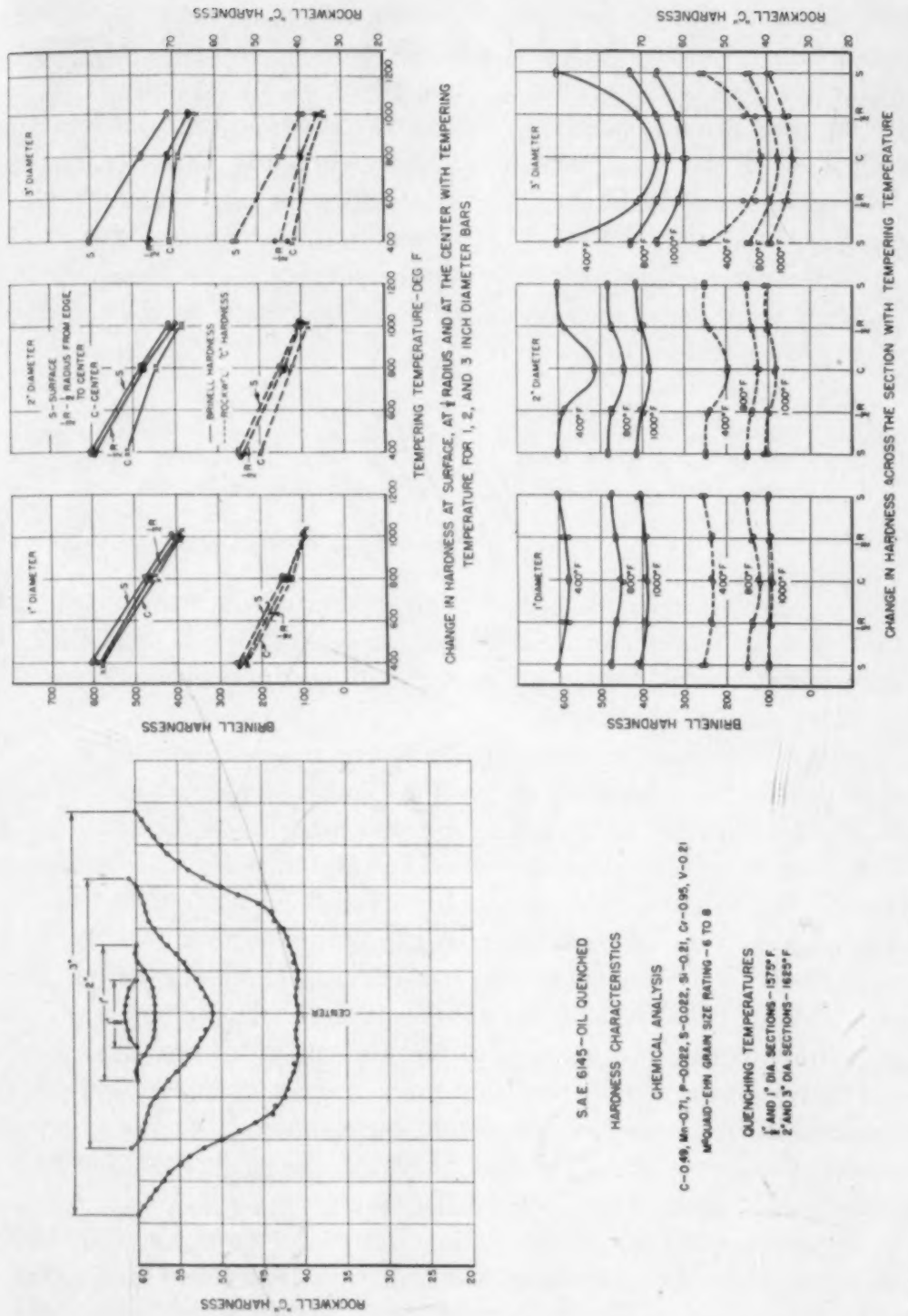
ROCKWELL "C" HARDNESS
3" DIAMETER
2" DIAMETER
1" DIAMETER
S - SURFACE
½R - ½ RADIUS FROM EDGE TO CENTER
C - CENTER
BRINELL HARDNESS
ROCKWELL "C" HARDNESS
TEMPERING TEMPERATURE - DEG F
CHANGE IN HARDNESS AT SURFACE, AT ½ RADIUS AND AT THE CENTER WITH TEMPERING TEMPERATURE FOR 1, 2, AND 3 INCH DIAMETER BARS
400°F
800°F
1000°F
BRINELL HARDNESS
CHANGE IN HARDNESS ACROSS THE SECTION WITH TEMPERING TEMPERATURE
CENTER
ROCKWELL "C" HARDNESS
S.A.E. 6145 - OIL QUENCHED
HARDNESS CHARACTERISTICS
CHEMICAL ANALYSIS
C-0.49, Mn-0.71, P-0.022, S-0.022, Si-0.21, Cr-0.95, V-0.21
McQUAID-EHN GRAIN SIZE RATING - 6 TO 8
QUENCHING TEMPERATURES
½" AND 1" DIA. SECTIONS - 1575° F
2" AND 3" DIA. SECTIONS - 1625° F

Fig. 20—Depth Hardness Characteristics of Oil-Quenched S.A.E. 6145 Steel.

Most of the ½-inch bars show an abrupt decrease in hardness from the surface to $\frac{1}{16}$ inch below the surface and a slight increase in hardness near the center. While tilting of the specimens during testing may often account for similar variations in hardness near the surface, for the specimens in question this was probably not the case. To demonstrate that the variations were real a ½-inch diameter specimen of S.A.E. 4130 steel was ground with two flats to the first transverse value $\frac{1}{16}$ inch below the surface. Ten Rockwell "C" readings were then taken on each flat. The results of this hardness survey are given as follows:

| Surface Hardness Mode of 40 Values | Transverse Value, $\frac{1}{16}$ Inch from Surface | Hardness Values on Flats Ground to First Transverse Values $\frac{1}{16}$ Inch Below Surface |
|---|---|---|
| 55 | 52, 51, 52, 52 | (1) 51, 52, 52, 52, 52, 52, 52, 52, 52, 52 |
| | | (1) 50, 51, 51, 51, 51, 51, 51, 51, 51, 51 |

The contours of the curves at the center of the water-hardening steels are quite regular, except for the 3-inch bar of S.A.E. 2330 steel which shows a marked increase in hardness near the center, probably due to segregation.

In order to show more clearly the relative differences in hardness penetrations between the various steels, a plot of their center hardness values for each section size was made, as shown in Fig. 21. The average center values were used for all the steels with the exception of the value for the 3-inch bar of S.A.E. 2330 steel. In this case a smooth curve was drawn at the center portion and the value selected was approximated from this curve. The plain carbon steels and S.A.E. T-1330 steel are represented by dashed curves. The carbon content of the latter is in the high range of the specification and therefore the steel is probably characterized by higher and deeper penetration of hardness than would be normal. Of this group of alloyed steels, S.A.E. 2330 and 5130 are the shallowest hardening while S.A.E. 6130 is the deepest hardening.

*Hardenability of Oil-Quenched Steels*—Maximum surface hardness was attained in all section sizes of the oil-hardened steels (Figs. 13 to 20) indicating that the critical cooling velocity had been equalled or exceeded. The pressure quench in oil was comparatively drastic as shown by the fact that published information on the hardenability of several of these steels quenched in oil by other methods failed to show maximum surface hardness even in 2-inch diameter sections.

From Figs. 13 to 20 it is observed that depth hardness penetration comparisons can be made of the oil-hardened steels only on the basis of the larger section sizes, since most of the ½ and 1-inch diameter bars were fully hardened and, in some cases, even the 2-inch bars were fully hardened.

A more convenient method of comparing the hardenabilities of the oil-quenched steels is also shown in Fig. 21, where the center

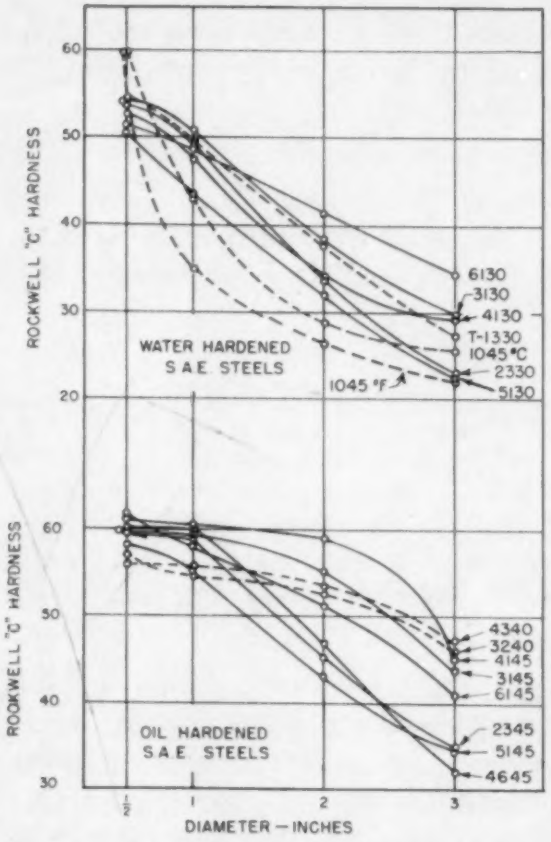


Fig. 21—Center Hardness Versus Diameter.

hardness values are plotted against the section diameter. The S.A.E. 3240 and 4340 steels appear to be deepest hardening. Of those containing 0.45 per cent carbon S.A.E. 3145 and 4145 steels are the deepest hardening while S.A.E. 2345, 4645, and 5145 steels are the shallowest hardening. S.A.E. 6145 steel falls in an intermediate hardening class by itself.

## Discussion of Tempering Data

*Water-Quenched Steels*—It was pointed out that the Brinell values for all the water-quenched and tempered pieces were obtained

using a tungsten carbide ball. The Brinell values of the tempered steels are shown in the upper or full-line curves while the Rockwell "C" values are shown in the lower or dash-line curves of Figs. 5 to 12.

The surface hardness of the S.A.E. 1045 fine- and coarse-grained types of steel was the same at each tempering temperature. The hardness values after tempering at the center and at half radius locations are higher for the coarse-grained steel, however.

S.A.E. 2330 steel (Fig. 7) responded to tempering more readily than other steels of the group. On the other hand, S.A.E. 6130 steel (Fig. 12) resisted softening to a marked degree. The remaining alloyed steels were intermediate in their response to tempering.

The following summary lists the surface hardness of S.A.E. 1045, 2330, and 6130 steels after tempering at 800, 1000, and 1200 degrees Fahr. and shows the relative resistance of the steels to softening on tempering.

| | Brinell Hardness Tempering Temp. Degrees Fahr. | | | Rockwell "C" Hardness Tempering Temp. Degrees Fahr. | | |
|---|---|---|---|---|---|---|
| | 800 | 1000 | 1200 | 800 | 1000 | 1200 |
| S.A.E. 1045 | 390 | 300 | 230 | 40 | 30 | 20 |
| S.A.E. 2330 | 340 | 270 | 220 | 33 | 27 | 17 |
| S.A.E. 6130 | 420 | 370 | 330 | 42 | 37 | 33 |

*Oil-Quenched Steels*—Figs. 13 to 20 contain the results of tempering experiments on the oil-quenched steels. The Brinell values of S.A.E. steels 4145, 5145, and 6145 were obtained using a tungsten carbide ball so that in the high range of hardness the values are expected to be somewhat higher than they would have been had a steel ball been used.

In this group of steels, S.A.E. 2345 steel softened most readily on tempering. Its approximate Rockwell and Brinell surface hardness values after tempering at the three temperatures were respectively, 54 and 540 at 400 degrees Fahr., 41 and 388 at 800 degrees Fahr., and 32 and 300 at 1000 degrees Fahr. Several steels of the group were somewhat more resistant to softening on tempering and had Rockwell and Brinell surface hardness values after tempering of 55 and 570 at 400 degrees Fahr., 45 and 450 at 800 degrees Fahr., and 36 and 350 at 1000 degrees Fahr. The S.A.E. steels 4145, 4340, and 6145 were least responsive to the tempering treatment.

ACKNOWLEDGMENT

The authors wish to acknowledge the contributions of the various members of the Sub-Committee on Alloying Elements to the planning of the conduct of this investigation.

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4. S. N. Petrenko, W. Ramberg, and B. Wilson: "Determination of the Brinell Number of Metals," National Bureau of Standards, Research Paper No. 903, Vol. 17, 1936, p. 59-95.

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## DISCUSSION

**Written Discussion:** By Gordon T. Williams, metallurgist, Testing and Research Laboratories, Deere & Co., Moline, Ill.

The authors are certainly to be thanked for these fine data. The thorough and painstaking manner in which it was undertaken and presented will make the paper very useful to user and investigator alike. Anyone who has performed such tests will realize the enormous amount of work involved, and will recognize the precision of the results.

Certain factors might well be mentioned in connection herewith. First, a holding time of 30-40 minutes at heat was used; as the present discusser shows in his paper at this convention, different heats of alloy steels vary in time necessary to attain maximum hardenability, and consequently some of the curves in this paper might have been raised in the center had longer times been used. The quench employed by the authors was excellent, far above that usually had in practice; this should be borne in mind. Also, it is well to realize that the special modes of transformation of some alloy steels (as shown by Griffiths, Pfeil, and Allen, Second Report of the Alloy Steels Research Committee, Iron and Steel Institute, 1939, Sec. XII, p. 343) mean that the ductility properties usually to be expected to accompany a given hardness may not be found.

If end-quench hardenability tests were or could be run on the steels here reported, such data would be a worthwhile addition to this already very valuable paper.

**Written Discussion:** By M. J. R. Morris, chief metallurgical engineer, Republic Steel Corp., Massillon, Ohio.

We are indebted to the authors for data of practical importance, although the maximum hardness values reported may not be realized in many heat

treating shops due to lack of oil or water flow as compared to that used by the authors.

The importance of carbon content on maximum hardness is indicated by the lower surface values for S.A.E. 3240 and 4340, but the higher alloy of these two grades begins to show its effect in the 3-inch round.

There appears to be some variation in the type of curve shown for the hardness penetration of the ½-inch rounds. Most of the curves for this size record a slight drop in hardness toward the center or uniform hardness across the diameter, but a few show an increase at the center. Were these bars checked for carbon segregation?

The degree of hardness penetration resulting from various alloys or combinations of alloy additions follows in general data accumulated from our past experiences, but the compilation of hardness values obtained on a series of tests all conducted under carefully supervised conditions is valuable for comparative purposes.

**Written Discussion:** By E. S. Rowland, research metallurgist, Steel and Tube Division, The Timken Roller Bearing Co., Canton, Ohio.

The authors are certainly to be commended for an excellent and painstaking piece of work. The data included form an important addition to the body of published information on hardenability of commercial steels.

We were interested to learn that the authors were unable to select proper quenching temperatures from observation of fractures as our own experience confirms this contention. Many metallurgists, however, seem to believe that selection of hardening temperatures of these steels may be made satisfactorily upon this basis.

The quenching fixture devised by the authors is certainly not the simplest one that could be used, but since such uniform hardness results were obtained with it, one cannot offer much in the way of criticism.

Since the authors failed to do so, the writer would like to inject a note of caution as to the use of their results in regard to commercial applications of some of these steels, particularly in large sizes. The surface and center hardnesses reported in this paper are the maximum obtainable and are not representative of even good commercial practice. For example, the authors show the center hardness of 4145 to be above that of S.A.E. 4340 in the 2-inch size. This is obviously due to the fact that both were quenched entirely through with their quenching method and the effect of carbon alone is important. In commercial practice, however, we may ordinarily expect S.A.E. 4340 steel to have a higher center hardness than S.A.E. 4145 in 2-inch round. It seems important, then, to remember that these data represent what can be obtained with these steels but are higher than what one may ordinarily expect to get in commercial practice.

**Written Discussion:** By B. R. Queneau, instructor, School of Mines, Columbia University, New York.

The authors have obtained a large amount of data on sixteen steels which should be of interest to many steel users. It is, however, unfortunate that no attempt was made by the authors to express their hardenability results in any other manner than by the well known hardness penetration curves.

At the Symposium on Hardenability at the 1938 American Society for Metals Convention, several attempts were made to present hardenability results in a quantitative manner. It would have been interesting if the authors had determined the severity of their quench and obtained the "H value" (heat transfer equivalent) for their quenching procedure. They could then have given the "critical size" of the various steels if quenched in an "ideal" medium as suggested by Grossmann, Asimow and Urban. On the other hand they could have used the Jominy Hardenability test, the Hardenability line, or even the S.A.C. rating as suggested by Burns, Moore and Archer, although in the latter case a size larger than 1 inch would have had to be selected as standard.

Below I have prepared a table of five steels of similar chemical analysis showing the "critical size" obtained by three different laboratories. It is interesting to note that three S.A.E. 1045 steels tested in three different laboratories all have a "critical size" of 7/8 inch. Although the authors state that their coarse-grained steel is definitely deeper hardening than their fine-grained steel, the difference in "critical size" was found to be only 1/8 inch. No grain size figures are given by either the authors or Burns, Moore and Archer for the steels at the quenching temperatures, so it is difficult to draw any definite conclusions. However, it can be seen that a considerable variation in hardenability can occur in steels of similar chemical analysis when made by different open-hearth practice.

| Steel | C | Mn | P | S | Si | McQuaid-Ehn Grain Size 1700 Degrees Fahr. | Fracture G.S. 1525 Degrees Cent. | Quenching Temperature Degrees Fahr. | Critical Size Inches |
|---|---|---|---|---|---|---|---|---|---|
| Klain and Lorig | | | | | | | | | |
| S.A.E. 1045-C | 0.44 | 0.75 | 0.019 | 0.025 | 0.25 | Coarse | ... | 1475–1525 | 1 1/8 |
| S.A.E. 1045-F | 0.44 | 0.72 | 0.018 | 0.032 | 0.20 | Fine | ... | 1475–1525 | 7/8 |
| Burns, Moore, Archer | | | | | | | | | |
| S.A.E. 1045-F | 0.48 | 0.78 | 0.016 | 0.030 | 0.27 | Fine | ... | ..... | 7/8 |
| Queneau and Mayo | | | | | | | | | |
| S.A.E. 1045-C | 0.41 | 0.82 | 0.028 | 0.032 | 0.23 | 2–4 | 5½ | 1525 | 7/8 |
| S.A.E. 1045-F | 0.41 | 0.81 | 0.015 | 0.028 | 0.15 | 6–8 | 7½ | 1525 | ½ |

**Written Discussion:** By W. E. Jominy, research laboratories division, General Motors Corp., Detroit.

The Subcommittee on Alloying Elements of the American Society for Metals Handbook Committee suggested that we apply the method of end quenching to these steels studied by Klain and Lorig.

This suggestion of the Subcommittee appealed to us as a means of adding to our fund of information on this test as well as to establish whether it is possible to convert the values from the one test to the other.

The method of measuring hardenability by end quenching consists in heating a bar 3 inches long and 1 inch round of the steel to be tested to a proper hardening temperature and then cooling the bar by spraying water on the bottom face only. The distance from the water-cooled end at which a given hardness is obtained is a measure of the hardenability of the steel.

In general, there has been an accumulation of data on the type of test used by Klain and Lorig and the question that arises in using the end quench-

ing method is whether the results of the end quenched bar can be converted to the older method of sectioning and measuring the hardness traverse.

Klain and Lorig sent us samples of each of the steels studied in this paper. All the steels were available in 1-inch round size except the S.A.E. 1045 fine-grained and the S.A.E. 4145 steels which were forged from 2-inch round to slightly over 1-inch round and turned to 1-inch round.

Of the steels received we have been able to complete our tests on ten which include S.A.E. 4340, 3240, 2330, T-1330, 3130, 4130, 5130, 6130, 1045F and 1045C. Two tests have been made on each of these steels using the plain end bar for one test and the cupped end bar for the other. These types of test bars have been described in a previous publication. It was found, as had been our early experience, that practically identical results are obtained with either of these test bars. We commonly obtain Rockwell readings within about 1 point C at the same locations with duplicate tests.

The heating temperature for these tests was the highest temperature listed in Table I for the steel except for steel T-1330 which was 1600 degrees Fahr. The total time in the furnace for all bars was 55 minutes.

In comparing the data from our tests with that of Klain and Lorig it is apparent that the quench employed by Klain and Lorig was quite severe. We have found that the location on our bar which compares with the center

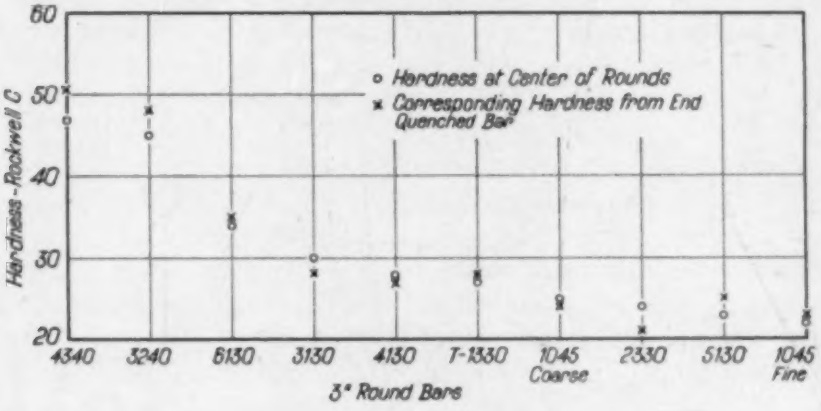



Fig. 1

of the 3-inch round bar quenched in oil was $1\frac{1}{16}$ inches, and for the 3-inch round quenched in water was ¾ inch. For the center of oil-quenched rounds the corresponding location on our bar for the 2-inch round was $\frac{11}{16}$ inch, for the 1-inch round was $\frac{5}{16}$ inch, and for the ½-inch round was ⅛ inch. For the center of the water-quenched bars the corresponding location on our bar for the 2-inch round was ½ inch, for the 1-inch round was $\frac{3}{16}$ inch and for the ½-inch round was $\frac{1}{16}$ inch.

In order to compare the results obtained by the two methods, Figs. 1 to 4 have been plotted. On these figures the steels have been arranged in the order of their center hardness as determined by Klain and Lorig. For each steel there is also plotted the center hardness indicated from the end quenched specimen.

Fig. 1 contains the steel in the 3-inch round size. The center hardnesses as determined by Klain and Lorig are indicated by the circles and the cor-

responding hardnesses from our bars are indicated by the crosses. It will be observed that quite good agreement exists between the two methods, especially when it is remembered that it is not uncommon in the method of Klain and Lorig to find variations in hardness for similar locations in the same bar of 3 points Rockwell C and occasionally as much as 7 points Rockwell C. Furthermore there no doubt were varying amounts and kinds of scale on these bars as they were placed in the quenching fixture by Klain and Lorig especially

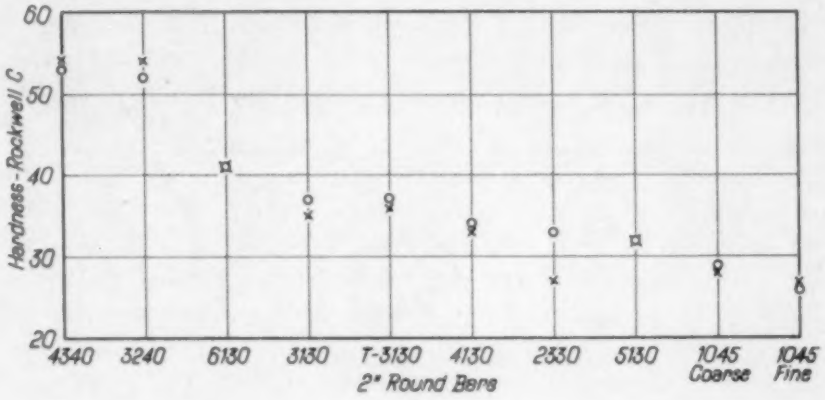



Fig. 2

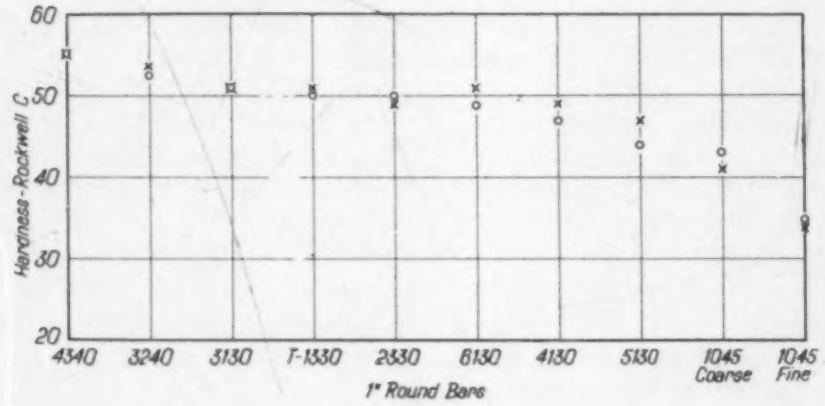



Fig. 3

since no precaution was taken to have a uniform finish on the bars. The variation in amount of scale undoubtedly affects the rate of cooling—especially with oil-quenching. In our end quenched bars there is practically no scale and the water-cooled surface always has the same finish.

Fig. 2 contains the same steels in the order of the center hardness of the 2-inch round bars. Here again there is good agreement except for S.A.E. 2330 steel where there is a difference of 7 points. We do not have any factual information as to why this difference should occur. Our first guess was that segregation existed but this was not found true. A suggested explanation is that a low heating temperature and resistance to scaling resulted in less scale on the S.A.E. 2330 steel thereby causing a relatively faster quench than on the other steels.

Fig. 3 shows good agreement between the two types of test for the 1-inch round bars and Fig. 4 likewise presents a favorable comparison for the ½-inch round bars.

It appears that we are justified in concluding that there is good agreement between the center hardnesses of various sized bars as obtained by the transverse method of measurement and the hardness at the point of corresponding cooling rate on the end quenched test bar.

One point of interest is that taking center hardness as a measure of hardenability it will be observed that these steels do not have the same relative hardenability in the 2-inch round size as in the 3-inch round size or in the 1-inch or ½-inch size. Each size seems to rate the steels in different order.

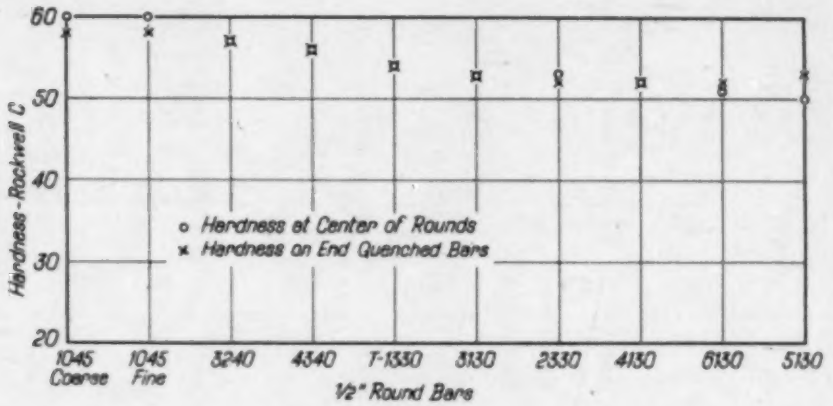



Fig. 4

However, if these steels are plotted according to expected center hardness from the end quenched specimens, they will have exactly the same order of relative hardenability in the 3-inch and 2-inch sizes, and except for transposing of S.A.E. 2330 steel from 10th to 7th place they have the same order of hardenability in the 1-inch round size. In the ½-inch round transposed hardnesses the effect of carbon content becomes rather strong and the same order of hardenability is not maintained.

The authors deserve credit for their lucid presentation of this large volume of data.

## Authors' Reply

The authors would like to extend their thanks and appreciation to the various discussers for the added information and for the clarification of some points.

We regret the insufficient emphasis on the severity of the quench used. As pointed out by the discussers, the quench was quite drastic and much higher than that ordinarily attained in commercial practice. In order to give an idea of the severity of quench used with respect to other quenching practices, the H-values for several of them are listed below, as given by Dr. E. C. Bain (Functions of the Alloying Elements in Steel, A.S.M. Educational Lectures, 1939, p. 170):

| Mode of Quench | H-Value |
|---|---|
| Cooling in still air | 0.02 |
| Moderate motion in oil | 0.4–0.5 |
| No motion in water | 1.0 |
| High velocity oil | 1.5 |
| Moderate motion in warm brine | 2.0 |
| Moderate motion in water | 4.0–6.0 |
| High velocity water | 10 and up |

The H-value for our oil quench was found to be approximately 1.5-1.7. However, the H-value for the water quench could not be determined accurately but appeared to be close to 10. Thus, the comparisons with H-values of other quenches indicate that maximum quenching effects were obtained in our tests.

We are glad to have Mr. Williams' comments regarding the effect of holding time and of quenching temperature on maximum hardenability obtainable. While much has been written about these two factors, yet little data have been presented. Further investigation is certainly needed, particularly for steels containing difficultly soluble carbides. His comment on the ductility of transformation products as observed by the English metallurgist is timely, since the effect of definite structures on tensile properties of the S.A.E. steels is contemplated for future investigation.

In answer to Mr. Morris' question, the ½-inch bars were not checked for carbon segregation. However, the increase in hardness from surface to center could also be due to retained austenite near the surface or to the presence of intermediate transformation products.

The analysis of the discrepancy in the 2-inch bars of S.A.E. 4145 and 4340 steels by Mr. Rowland is quite correct. We are glad to learn that our inability to predict the proper quenching temperatures from observations of fractures is in line with the experiences of Mr. Rowland.

We are sorry that additional quantitative relationships were not included but to do so using the various methods suggested by Dr. Queneau would easily have made another paper. However, Mr. Jominy's discussion supplements our data and partly answers Dr. Queneau's criticism. The comparison data on S.A.E. 1045 steels are much appreciated.

The authors are very grateful to Mr. Jominy for the additional data on the correlation of the end-quench with the traverse hardness method. We are pleased to observe such excellent agreement in the two methods. However, it is regretted that the end-quench method was not applied to some larger sizes to check the observation that each size in the traverse hardness method rated the steels in different order. It must be remembered that the end-quench results for the different sizes of any one steel are obtained from one specimen, whereas in the other method the results are obtained from different bars and thus many more variables are introduced during the processing.

# THE EFFECT OF HARDNESS AND OF TEMPERATURE ON THE STRENGTH, DUCTILITY, AND TOUGHNESS OF A HEAT TREATED CARBON STEEL

By S. W. Lyon

*Abstract*

*Tension tests, impact-tension and impact-bending tests were made on unnotched specimens of steel heat treated to produce a range of hardness. The results of these studies indicated that the steel loses strength, ductility and toughness quite markedly in the range of Rockwell hardness of C-47 to C-52. The effect of a testing temperature from + 70 degrees Fahr. (+ 21 degrees Cent.) to — 70 degrees Fahr. (— 57 degrees Cent.) on the toughness of the steel was also studied. A method for determining the energy correlation between full-size and fractional-width unnotched impact specimens is presented.*

## Introduction

IN connection with the application of a hardened carbon steel to a service application requiring resistance to impact and battering wear, a program of physical testing was carried out to determine the suitability of the material. The results of the preliminary tests brought out the somewhat surprising fact that, without exception, in the range of hardness considered (Rockwell C-30 to C-44) the hardened material not only developed higher tensile strength but also higher toughness than the original hot-rolled material. In the range of hardness considered, the toughness, especially, seemed to increase with increase in hardness. This rather unlooked for observation led to the question of how the material varied in strength and toughness over a range of Rockwell hardness from a maximum, obtainable by ordinary heat treatment procedure, to a minimum value equal to that of the original hot-rolled material. As the particular service application under consideration would impose a range of temperature conditions from + 70 degrees Fahr. (+ 21 degrees Cent.) to — 30 degrees Fahr. (— 34 degrees Cent.), information was also desired as to the effect of temperature on the toughness of the steel.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. The author, S. W. Lyon, is special research assistant in engineering materials, Engineering Experiment Station, University of Illinois, Urbana, Illinois. Manuscript received June 26, 1939.

The tests herein reported were carried out to provide answers to the above two questions. The physical properties of the heat treated steel studied and the methods used are briefly as follows: Several series of tension specimens of different hardness were prepared and tested to study the variation of tensile strength, percentage elongation, and percentage reduction of area. Impact-bending tests of unnotched bars and impact-tension tests were used to determine variation of toughness[1] with hardness changes. Impact-bending tests of unnotched specimens tested at temperatures varying from + 70 degrees Fahr. (+ 21 degrees Cent.) to — 70 degrees Fahr. (— 57 degrees Cent.) provided information on the effect of temperature on the toughness of the steel.

## Test Specimens

The specimens for test were machined from a hot-rolled bar whose heaviest section measured 1.5 x 3 inches. Slices from this heavy section were hot-etched in a 50 per cent solution of hydrochloric acid to determine that the steel for test was free from flakes, pipes and other macrographic defects. In order to insure, as far as possible, that the material should be of uniform quality, all specimens were cut from the bar adjacent to one rolled surface. The chemical analysis of the steel is as follows: Carbon 0.73 per cent, manganese 0.88 per cent, silicon 0.14 per cent, phosphorus 0.012 per cent, sulphur 0.022 per cent.

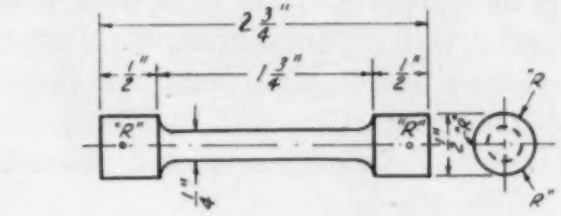
Fig. 1 shows the various types of test specimens used. The specimens were machined to within 0.010 inch of finished dimensions, heat treated to produce the desired hardness and then finish-ground to size. Rockwell "C" hardness readings at the locations shown in Fig. 1 were taken on all specimens prior to testing.

The proper heat treatment to be applied to the specimens had been determined in preliminary tests by varying quenching medium and temperature. The general procedure used was to quench all specimens to obtain maximum hardness, followed by suitable draw to obtain the desired hardness. For specimens of the small dimensions used it was found necessary to use an oil quench to prevent quenching bursts which were frequently met with when water was used as the medium. The combination of quenching temperature and medium chosen imparted maximum uniform hardness. The heat treating procedure which was used on all specimens was as

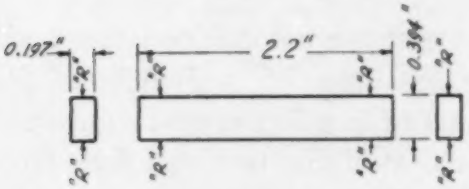
[1]The term "toughness" is used to denote energy required to fracture a metal.

follows: Heat to 1400 degrees Fahr. (760 degrees Cent.), hold 30 minutes, quench in oil; draw for one hour at temperature necessary to produce the desired final hardness.

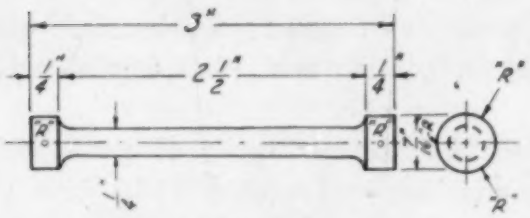
Fig. 2 is a group of four photomicrographs taken from specimens of different hardness as shown. The structure shown in (a) which is that of the as-rolled material consists entirely of pearlite.




TENSION SPECIMEN




IMPACT BENDING SPECIMEN




TENSION IMPACT SPECIMEN

NOTE: "R" INDICATES LOCATION OF ROCKWELL "C" HARDNESS DETERMINATIONS

**Fig. 1—Various Types of Test Specimens Used in this Investigation.**

A specimen with the hardness of C-48 developed the transition structure (c) which falls between the troostite of (b) obtained for the material of hardness C-64 and the sorbitic structure (d) shown for the material of hardness C-40.

## Test Methods

The tension test specimens which were prepared and tested to study the effect of variation of Rockwell "C" hardness on the tensile strength, percentage elongation, and percentage reduction of area

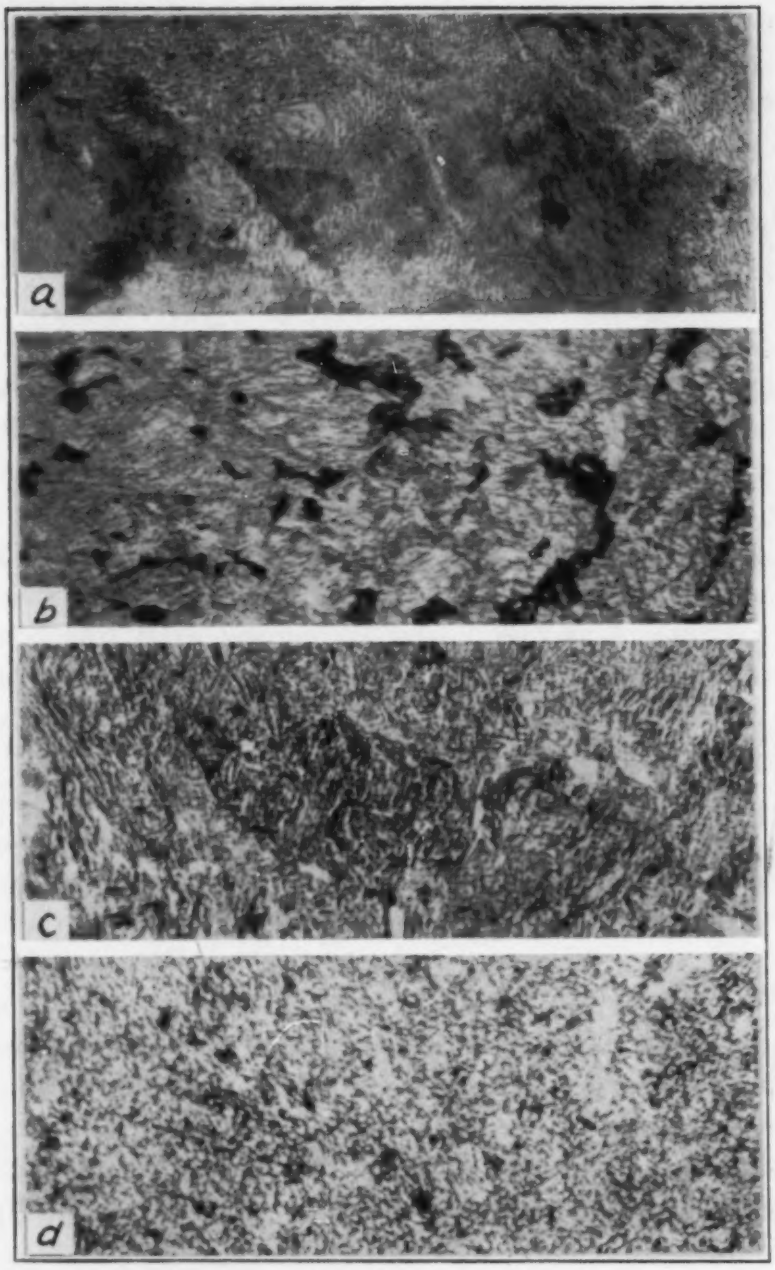



Fig. 2—Photomicrographs Showing the Effect of Heat Treatment. Etched in 2 Per Cent Nital. × 600.

a. As-Rolled Steel .....Rockwell "C" Hardness 25
b. Heat Treated Steel .....Rockwell "C" Hardness 64
c. Heat Treated Steel .....Rockwell "C" Hardness 48
d. Heat Treated Steel .....Rockwell "C" Hardness 40

were tested in a 3-screw, 50,000-pound capacity universal testing machine. Robertson shackles, which tend to minimize eccentricity of loading, were used during the tests. A speed of head of testing machine of 0.05 inch per minute was employed for all tests.

Tension-impact tests were made on a 223-foot-pound capacity Charpy impact machine which was fitted with a special tension attachment. The outstanding features of this tension attachment were the spherical seats incorporated in its design which tended to a great extent to eliminate bending of the specimens during test. Unnotched specimens were used as it was felt that energy values obtained from such specimens would be more indicative of the toughness of the steel than if notched ones were used. As for the tension tests, several series of specimens were prepared, heat treated, and tested to cover the range of hardness studied. The results obtained were considered as a measure of the variation of the toughness of the steel with variation of hardness.

The unnotched impact bending specimens were tested in bending, using the Charpy machine. For these tests the tension attachment was removed and the machine was employed as for conventional Charpy testing. Unnotched bars were used as it was felt that by elimination of the notch specified in the standard Charpy specimen, a better measure of the toughness of the steel could be obtained. It was found during the preliminary testing of the unnotched bending specimens of 0.394 inch by 0.394 inch cross section in the Charpy machine that the heat treatment had imparted to some of the series of specimens falling within a certain range of hardness, impact resistance in excess to the testing machine capacity. (The impact resistance was so great as to completely stop the pendulum at the point of impact.) To complete the toughness-hardness correlation curve it therefore became necessary to determine energy values for steel so tough that an unnotched specimen of standard cross section could not be broken in the 223-foot-pound capacity Charpy machine. The method which proved most satisfactory for the determination of such high energy impact values was, briefly, that of testing series of fractional width specimens and establishing a correlation between the energy values obtained for these and for the standard size specimens. (A detailed description of this method of determining the energy correlation between full size and fractional width unnotched impact specimens is given in Appendix (A) of this paper.) Fig. 1-C shows the type of fractional width specimen used to obtain the impact-bending versus Rockwell "C" hardness relationship curve.

To determine the variation of toughness of the steel with change of temperature, the energy developed in impact-bending tests of unnotched specimens was considered as the criterion of tough-

ness. The variation of toughness with change of temperature was obtained by determining the impact-bending results versus Rockwell "C" hardness relationships for five testing temperatures, namely $+70$, $+30$, 0, $-30$, and $-70$ degrees Fahr. The method used to cool the specimens for testing at the various lowered temperatures was as follows: An insulated container of sufficient capacity to hold a 1500 cubic centimeter bath of acetone was provided. The acetone bath was then cooled to the desired temperature by the addition of pieces of dry ice (solid $CO_2$). The temperature of the bath was measured by a standard A.S.T.M. low-temperature thermometer immersed in the acetone. It was found that after the specimens were immersed in the coolant the temperature of the bath could be adjusted and held to within a limit of closer than plus or minus one degree Fahr. When the test specimens had remained in the bath long enough (3 to 5 minutes) to ensure their being completely cooled to the temperature desired, they were removed one at a time with rubber-tipped tongs, quickly placed in the testing machine and tested. Preliminary tests indicated that a specimen could be tested within 10 seconds of removal from the bath with the loss of but one degree of temperature. The average time for the interval between testing and removal from the bath for the tests herein reported was 3 seconds; for this time interval no temperature change was discernible.

In order to provide a comparison between the variation of impact strength of heat treated steel and as-rolled steel with change of temperature, four or more specimens of the as-rolled material were tested at each of the five testing temperatures.

## Results of Tests

Fig. 3 gives the results obtained from the tension tests. Examination of the curves shows that maximum values of tensile strength were obtained at the hardness of approximately Rockwell C-52 and that for hardnesses in excess of this value the tensile strength values fell off rapidly. These test results, then, seem to indicate that from the standpoint of tensile strength the usable hardness for this steel is exceeded when a value is obtained higher than approximately Rockwell C-52. Values for reduction of area also fell off rapidly when the hardness of C-52 was exceeded, while elongation values decreased more gradually.

The results of the impact-bending and impact-tension tests of

unnotched test specimens of varying hardness are given in Fig. 4a and 4b. Considering these test results as criterion of toughness it can be observed that the toughness of this heat treated steel drops off rapidly when the hardness of Rockwell C-48 is exceeded. Impact-bending test results obtained by Davenport, Roff and Bain[2] on steel having a carbon content similar to that of the steel used in the test

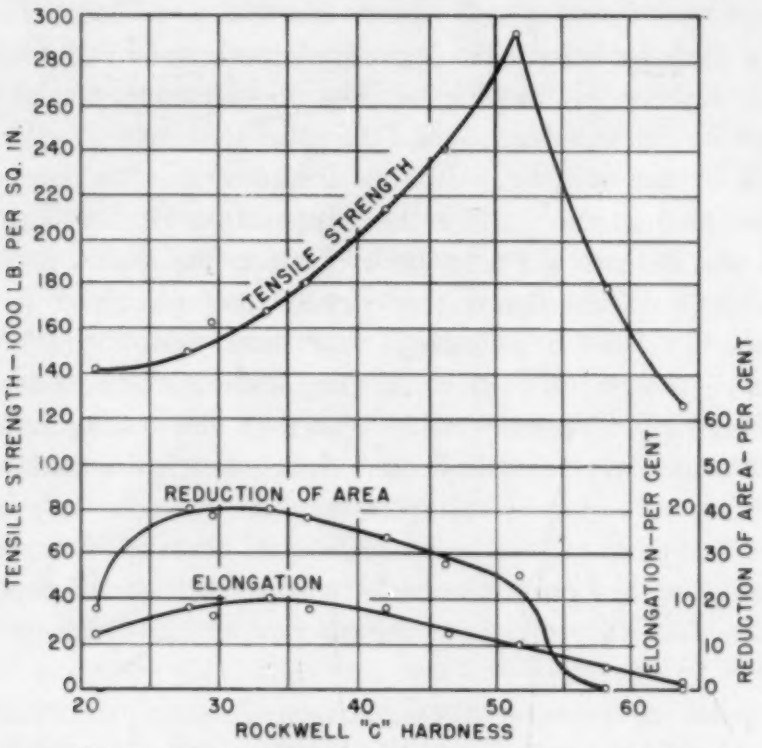



Fig. 3—Tensile Strength, Reduction of Area, and Elongation Versus Rockwell "C" Hardness.

herein reported, also indicated that the impact resistance or toughness fell off very rapidly when the hardness of the steel exceeded the value of Rockwell C-50. Further examination of the curves given in Fig. 4 will show that the plotted test results from some specimens of similar hardness, i.e., C-48, group themselves on the upper and lower levels of the impact strength curves. This indicates that there is a very abrupt transition change in the property of the material to resist impact loads. This abrupt change of impact strength is apparently accompanied by some instability as evidenced by the spread of impact strength for specimens of similar hardness.

Fig. 5 gives the "Rockwell 'C' hardness versus Charpy impact strength relationships" obtained for the five testing temperatures.

[2]E. S. Davenport, E. L. Roff and E. C. Bain; "Microscopic Cracks in Steel," TRANSACTIONS, American Society for Metals, Vol. XXII, No. 4, 1934, p. 297, Fig. 5.

It will be observed that the test values obtained for the as-rolled specimens are included and appear in the lower left corner of the diagram. The envelope was constructed by superimposing the hardness versus impact relationship curves for the various testing temperatures and then enclosing all five curves. For clearness the individual superimposed curves have been omitted and the resulting

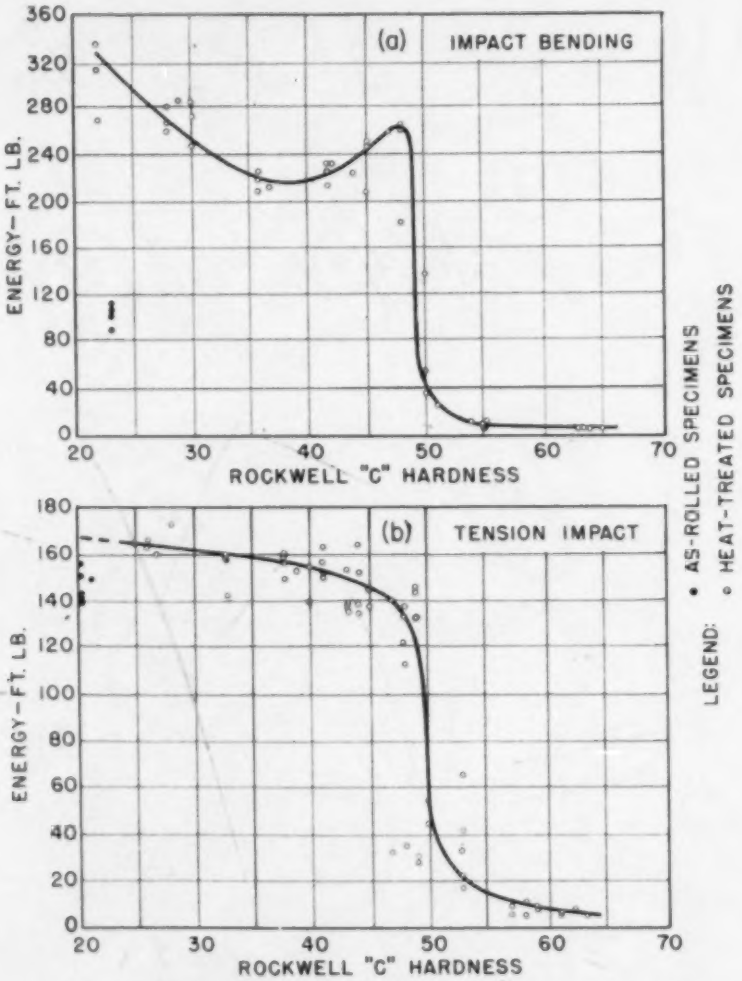



Fig. 4—Impact Test Results Versus Rockwell "C" Hardness.

cross-hatched envelope gives the spread of average values of hardness versus Charpy impact strength obtained for the range of testing temperature between + 70 degrees Fahr. and — 70 degrees Fahr. It will be noticed that the envelope has the characteristic form of the individual curves with a very sharp drop off in impact strength occurring at all temperatures when the hardness range of C-47 to C-50 is exceeded. These results then indicate that the toughness of this heat treated steel is not greatly changed or affected by tempera-

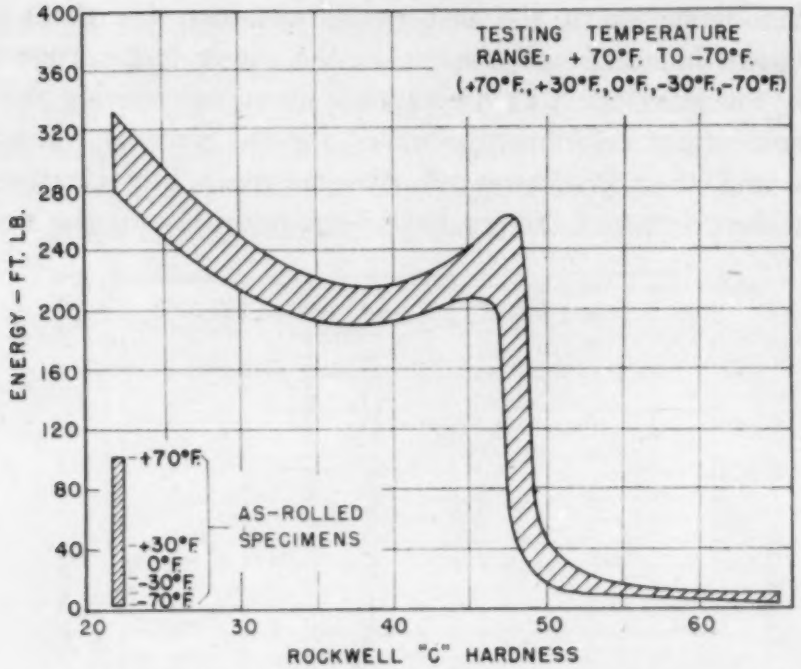



Fig. 5—Envelope Showing Variation of Rockwell "C" Hardness Versus Charpy Impact Strength Relationships Caused by Varying Testing Temperatures Between + 70 Degrees Fahr and — 70 Degrees Fahr.

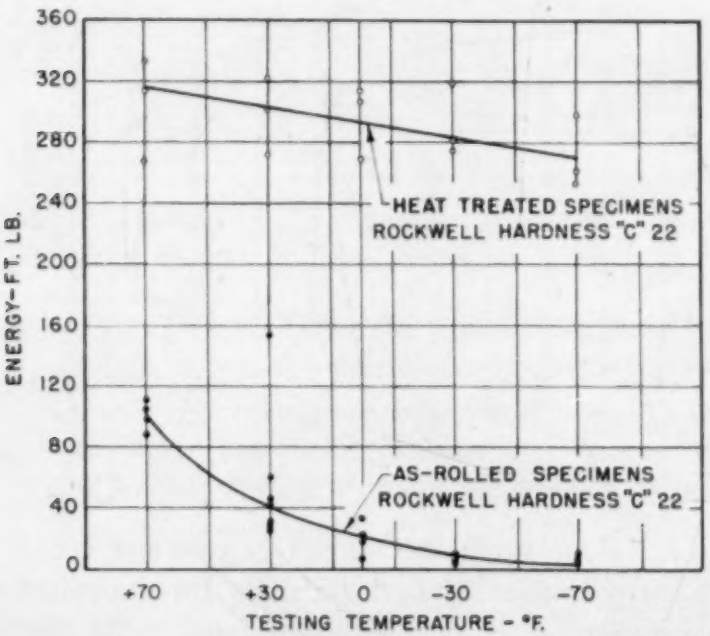



Fig. 6—Comparison of Variation of Charpy Impact Strength of Heat Treated and As-Rolled Specimens of Steel Tested Over a Range of Temperature of +70 Degrees Fahr. to —70 Degrees Fahr.

tures varying between + 70 degrees Fahr. and — 70 degrees Fahr. Also, for temperatures between + 70 degrees Fahr. and — 70 de-

grees Fahr. it is indicated that to have high impact resistance this heat treated steel should not exceed C-47 in hardness.

As previously noted the impact-bending test results of the as-rolled specimens are given at the lower left corner of Fig. 6. The average value of impact strength of as-rolled steel falls off from 100 foot-pounds at + 70 degrees Fahr. to less than 5 foot-pounds at — 70 degrees Fahr. Also, the change of temperature of testing from + 70 degrees Fahr. to + 30 degrees Fahr. resulted in a loss of strength from 100 to 42 foot-pounds. Fig. 6 presents the above results in another form and in addition for comparison presents the results of the impact tests at the five testing temperatures of the heat treated specimens of the same hardness (C-22) as the as-rolled specimens. The heat treated specimens have a high impact resistance and over the range of testing temperatures lose but 15 per cent of their strength against 95 per cent loss for the as-rolled specimens. Also, it should be noted that for equal hardness the heat treated specimens at + 70 degrees Fahr. have an average energy value of 290 foot-pounds against 100 foot-pounds for the as-rolled specimens. These results indicate then that the heat treatment of the steel tested produced two beneficial results, namely, the impact strength was much improved over the as-rolled condition and that lowering of temperature had much less effect on the impact strength.

## Summary of Conclusions

1. The tension test results, the impact-bending and impact-tension test results and impact-bending results for a temperature range of + 70 to — 70 degrees Fahr. all show that the heat treated steel tested loses strength, ductility, and toughness quite markedly in the range of Rockwell hardness of C-47 to C-52.

2. Conclusion 1 then would indicate that the *usable* hardness of the heat treated steel is exceeded when a hardness of approximately C-47 is exceeded. In other words, this material could not be used in any service application where a hardness of Rockwell C-47 was exceeded without the considerable sacrifice of strength, ductility and toughness.

3. The results of the impact-bending and impact-tension tests show that the heat treated steel developed good toughness for hardnesses below C-47 at all testing temperatures. This would point out that if this heat treated material were used in a service application the lower limit of hardness would not be governed by toughness

considerations. Rather, wear or strength considerations would be the criterions for fixing the lower limit of hardness.

4. The toughness of the heat treated material over the range of hardness studied was not greatly affected by temperature variations between + 70 degrees Fahr. and — 70 degrees Fahr. as shown by the impact-bending results for that temperature range.

5. Comparison of the toughness of specimens of the as-rolled and heat treated steel of the same hardness indicates that the heat treatment produced the following beneficial results:

(a) Heat treatment improved the impact-bending resistance or toughness quite markedly.

(b) The heat treated steel was much less affected by a testing temperature range of + 70 degrees Fahr. to — 70 degrees Fahr. than was the as-rolled material.

## Appendix (A)

### A Method for Determining the Energy Correlation Between Full-Size and Fractional-Width Unnotched Impact Specimens.

In brief, the method used for determining the energy correlation between full-size and fractional-width unnotched impact specimens consisted of testing specimens of varying width in a 223-foot-pound capacity impact testing machine. By plotting the test results, a curve was obtained which gives a relationship between energy values and specimen size.

To carry out this program, eight series of test specimens were prepared from the steel previously described. Four of these series of specimens were tested in the as-rolled state and four series were heat treated to produce a toughness in the full-size specimens of approximately 200 foot-pounds or just within the machine capacity.

Each series consisted of fifteen specimens divided into three groups of five each with dimensions as shown in Fig. 7 (a, b, c). It should be noted that the width of specimen is the only dimension that varies for the three types a, b, and c. As all specimens were tested in bending in a Charpy impact machine, the type of loading can be represented schematically as shown in Fig. 7d. Consideration of static test relations suggests that when the width of specimen (B) only is varied, theoretically the resistance to fracture should vary directly as (B).

Fig. 8 gives the correlation obtained by plotting the test results

for the eight series of specimens tested. In order to provide a direct comparison between energy values obtained for the various series of specimens the test results are given on a percentage basis. For each series the average energy values obtained by test for the type (a), (0.394 by 0.394 inch) specimens, are assigned as being 100 per cent values. The average energy values obtained from each of the other two widths of specimens (b and c) was then converted to a

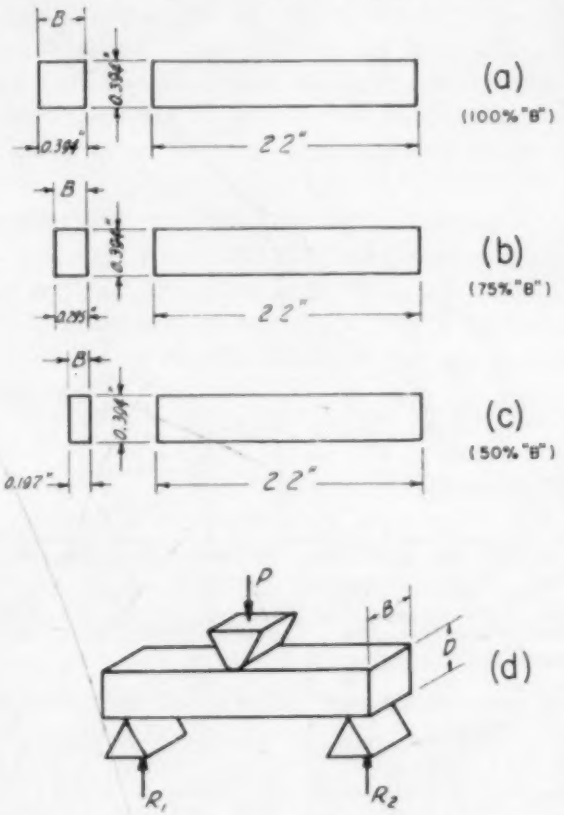



Fig. 7—Types of Unnotched Charpy Specimens.

percentage of that obtained for type (a). The curve of Fig. 8 was obtained by plotting per cent energy against per cent width (B).

In using the correlation curve of Fig. 8 in connection with subsequent tests, it was found convenient to derive conversion factors for the width of specimen in question so that these conversion factors when multiplied by the test values would give energy values equivalent to those which would have been obtained should full-width specimens have been used. Fig. 8 gives these conversion factors as 1.36 and 2.18 for specimen widths of 0.295 inch and 0.197 inch, respectively.

Considerable use of the above method of the evaluation of the energy of fracture of fractional-width impact-bending specimens for

the steel in question and also for other materials has tended to engender confidence in the reliability of this method. B. G. Aldridge and G. C. Shepherd, Jr.[3] reported the successful employment of a quite similar method of evaluating test data from fractional-width Charpy specimens. Further, it is believed that this method of deter-

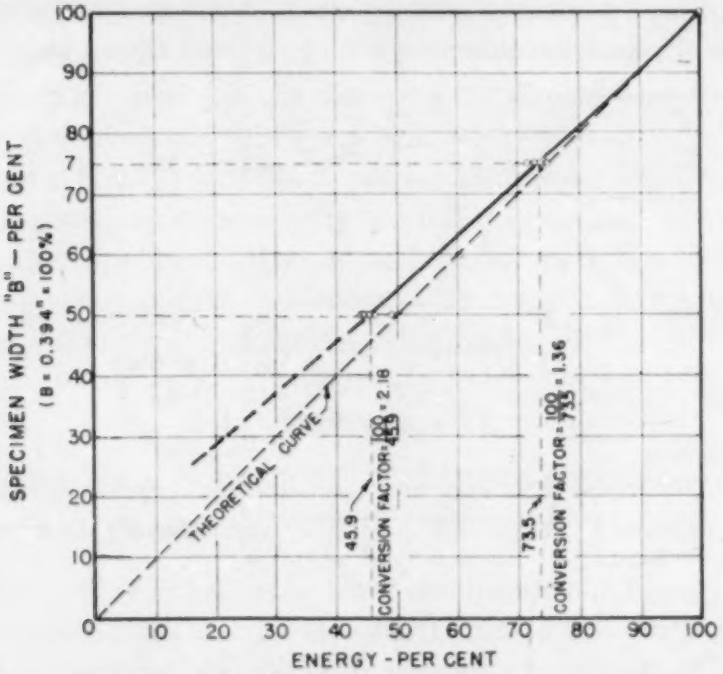



Fig. 8—Graph Showing Effect of Varying the Width of Unnotched Bars Tested in a Charpy Impact Testing Machine.

mining equivalent impact energy values for unnotched Charpy specimens of varying width provides a convenient means of determining impact values for unnotched bars of metals in many cases where it is impossible to obtain a full-width specimen, or where a full-width specimen could not be broken with the impact testing machine available.

## Acknowledgments

The work herein reported has been carried out under the supervision of Professors H. F. Moore and H. R. Thomas of the staff of the Engineering Experiment Station of the University of Illinois. J. S. Shapland and R. S. Hutton, student assistants, have helped materially with the testing and preparation of data. Acknowledgment is gratefully extended to the above named for their helpful criticism and assistance.

[3]B. G. Aldridge and G. C. Shepherd, Jr.; "Nickel Steels at Low Temperatures," *Metals and Alloys*, June 1936, Vol. 7, No. 6, p. 147-152.

## DISCUSSION

**Written Discussion:** By H. B. Wishart, metallurgical laboratory, Carnegie-Illinois Steel Corp., Gary, Ind.

The data contained therein clearly show the material benefit obtained through better bending impact values at both atmospheric and low temperatures when steel is heat treated to give a hardness between 20 and 47 Rockwell "C". It is interesting to note that a marked drop occurs in impact resistance when specimens are hardened over 47 Rockwell "C".

The impact bending data and the impact tension data give curves that are somewhat comparable as to shape. There is, however, a marked difference between the impact values of as-rolled specimens as compared to heat treated specimens of the same hardness when tested in bending impact. This difference is practically negligible when the same comparison is made on specimens tested in tension impact.

It is realized that the two types of impact testing are not directly comparable. It does, however, seem odd that heat treatment in one case has such a beneficial effect, while in the other, apparently little benefit is derived from such treatment. We would like to ask the author if he has an explanation for this phenomenon, and if not, has he observed this same condition in steels of other compositions?

### Author's Reply

The author is indebted to Mr. Wishart for his interesting discussion. In reply to the request for an explanation of why there is a marked difference between the impact values of as-rolled specimens as compared to heat treated specimens of the same hardness when tested on one hand in impact bending and on the other hand in impact tension (Fig. 4), it may be said that due to lack of experimental data any answer which the author may put forward will necessarily fall in the category of conjecture.

However, consideration of the mechanism of fracture under the conditions of the impact test can conceivably be divided into two phases, first, crack formation and secondly, crack propagation. It is also well to bear in mind that the bending impact test produces a much different type of failure from that of the tension impact test.

Failure of a specimen by impact bending occurs by first having the relatively few outermost fibres highly stressed in tension with a crack forming at this location; after the crack is formed the crack rapidly progresses to failure over the entire critically stressed cross section. In conformity with this picture of failure a relatively small portion of the energy of failure would be consumed in crack formation and the major portion in crack propagation.

On the other hand failure of a specimen by impact tension occurs by first having almost all of the fibres of the critically stressed cross section highly stressed in tension to the stage where crack formation occurs; then presumably many cracks form and crack propagation to failure is rapid indeed. In other words the major portion of the energy of fracture is used during the first phase of failure—crack formation—the remaining relatively small amount is absorbed in crack propagation.

To the author it is conceivable that the difference in impact values under discussion might be accounted for by the divergent conditions imposed by the bending impact test where the major portion of the energy of fracture was expended in crack propagation and the tension impact test where the major portion of the energy of fracture was expended in crack formation.

Indirect confirmatory evidence to support this contention is given by the tests of Collins and Dolan.[1] Tension impact and bending impact tests were made on notched and unnotched specimens of several steels. The impact bending results on unnotched specimens were not included as the materials were so ductile as to not break when tested in the unnotched condition. For steels B, C, and D the results in the following table were obtained. The significance

| Material Designation | Impact Tension Energy Ft. Lb. Notched | Impact Bending Energy Ft. Lb. Notched |
|---|---|---|
| B | 30.7 | 112.0 |
| C | 49.8 | 24.4 |
| D | 56.4 | 13.4 |

of these results is that the impact tension values vary but very little for the three different materials while the impact bending values show a considerable spread. Considering that the as-rolled steel and the heat treated steel of the paper under discussion (Lyon) are different materials as produced by heat treatment, then the difference in spread of impact values as given by the tension impact tests and bending impact tests is consistent with the results obtained by Collins and Dolan.

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[1]W. L. Collins and T. J. Dolan, "Physical Properties of Four Low-Alloy High-Strength Steels," *Proceedings*, American Society for Testing Materials, Vol. 38, 1938, Table IV, p. 162.

# THE ACCELERATING EFFECT OF CERTAIN METALLIC ELEMENTS ON GRAPHITIZATION

By H. A. Schwartz, Vincent Fiordalis, John L. Fisher, James F. Shumar and M. J. Trinter

*Abstract*

*This paper represents a restudy of the known graphitization accelerating effects of copper and nickel on white cast iron for comparison with a recent study of the graphitization retarders manganese, chromium, vanadium and molybdenum.*

*A major purpose of the investigation was an attempt to learn something of the mechanism by which metallic alloys alter graphitizing rates by studying accelerators as well as retarders.*

*Weight for weight the effectiveness of copper and nickel in promoting graphitization at 900 degrees Cent. (1650 degrees Fahr.) is found to be closely alike. One per cent of either alloy approximately doubles the graphitizing rate of an iron already containing about 0.90 per cent silicon. Neither element in amounts up to 1 per cent greatly affects the solubility of graphite in gamma iron. Copper has but little effect on the nodule number at 900 degrees Cent. (1650 degrees Fahr.) and nickel has but little on the number of large size nodules but increases the number of very small nodules. Either element increases the migratory rate of carbon in gamma iron.*

*The data at 700 degrees Cent. (1290 degrees Fahr.) are less conclusive. While both elements produce an approximately similar acceleration of graphitization, copper seems to be effective mainly by increasing the nodule number. The observations of this paper should be compared with an earlier paper.*[1]

The data on nickel are taken from a thesis by Messrs. Fiordalis and Trinter prepared in partial fulfillment of the degree of Bachelor of Science in Metallurgical Engineering at Case School of Applied Science and the data on copper are taken from a thesis by Messrs. Fisher and Shumar prepared in partial fulfillment of the degree of Bachelor of Science in Metallurgical Engineering at Case School of Applied Science.

[1]H. A. Schwartz, H. H. Johnson and C. H. Junge; "The Retarding Effect of Certain Metallic Elements on Graphitization," Transactions, American Society for Metals, Vol. 25, 1937, p. 609.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, H. A. Schwartz is manager of research, National Malleable and Steel Castings Co., Cleveland, and Messrs. Vincent Fiordalis, J. L. Fisher, J. F. Shumar and M. J. Trinter were students, Case School of Applied Science, Cleveland. Manuscript received May 19, 1939.

THE fact that nickel and copper accelerate graphitization is well known (1), (2).[2] It is here desired to add certain information suggested by more recent studies and to tie the action of these two accelerating metals to that of certain retarders recently studied (3). We wished to approach the subject in the same way and make records comparable in general with the observations of (3) all with the purpose of generalizing so far as possible our knowledge of the effect of alloying elements and the mechanism of that effect. Many useful explanatory references will be found in that paper and will not, in the interest of brevity, be here repeated.

## Experimental Procedure

The specimens were in the form of the usual tensile test specimens as described in A.S.T.M. specification A 47-33. The metal to which additions were made came, as did the first melted metal of the reference from a Brackelsberg furnace heat made at the Indianapolis Works of the National Malleable and Steel Castings Co. which had the composition:

| | Per Cent |
|---|---|
| Carbon | 2.42 |
| Silicon | 0.91 |
| Manganese | 0.30 |
| Sulphur | 0.105 |

By permission of S. C. Wasson, manager of that plant, ladle additions of nickel or copper as turnings in various amounts were made to a series of ladles and test specimens poured in green sand molds. It was intended to produce samples containing 0.25, 0.50, 0.75 and 1 per cent of the alloying metal, each specimen containing but one alloy. The actual recoveries were as follows:

| Nominal Alloy Content | Copper Series, Per Cent Copper | Nickel Series, Per Cent Nickel |
|---|---|---|
| 0.25 | 0.33 | 0.24 |
| 0.50 | 0.56 | 0.46 |
| 0.75 | 0.73 | 0.66 |
| 1.00 | 0.99 | 0.91 |

Portions of the grip ends of specimens of each alloy and of unalloyed metal were packed in pipe nipples in cast iron borings and heated for various times at 900 degrees Cent. (1650 degrees Fahr.)

[2]The figures appearing in parentheses refer to the bibliography appended to this paper.

in a "Hump" furnace equipped with automatic temperature control. The heating rate of the charge, approximately 1.5 hours to temperature, was determined on a dummy pot and used in determining the actual time at temperature. All the specimens in a given nipple were equidistant from the cylindrical surface. One series was also run at 700 degrees Cent. (1290 degrees Fahr.). After cooling, the specimens were analyzed for graphite and the nodule number (considered approximately constant) determined, by the commonly used method, usually on specimens from the longest heat treatment. The procedure is in all respects similar to that used in the investigation of the retarding elements when added in the ladle to fuel-melted iron.

## Observational Data

There seems little reason for elaborating on the observational data in detail; they are therefore condensed in the following tables.

The occasional negative values in the tables, having no physical meaning, are due to discrepancies between the existing conditions and the postulated distribution into five sizes of spheres. When the number of any one size is small it is possible to get fictitious negative values. Some intercepts have been assigned to the adjacent sizes which belonged to that size showing a negative value.

## Discussion of Results

It is already established (4), (5) that, except for a possible interference due to slow linear crystallization velocity of graphite, the graphite formed in a given metal at given temperature should be proportional to the 1.5 power of time until somewhat over one-half the available combined carbon is graphitized. If then one plots time versus graphite on log-log paper, the straight line having a slope of 3 in 2 which best passes through the points, below about 0.70 per cent graphite in the present case at 900 degrees Cent. (1650 degrees Fahr.) is the most probable locus of the time-graphite curve, and its intercept with the line of unit time (1 hour) marks the value of the coefficient of graphitizing rate.

On account of the (only apparent) increase of sensitivity at low graphite contents it is best to give most weight to the higher graphite points. Such graphs are shown in Fig. 1. The concordance

**Table I**

**Graphite Content of Copper-Bearing Iron in Relation to Time at 900 Degrees Cent. (1650 Degrees Fahr.)**

| Per Cent Copper | Per Cent Graphite Formed in Indicated Number of Hours at 900 Degrees Cent. (1650 Degrees Fahr.) | | | |
|---|---|---|---|---|
| | 2 Hours | 4 Hours | 6 Hours | 12 Hours |
| trace | 0.06 | 0.22 | 0.70 | 1.23 |
| 0.33 | 0.16 | 0.48 | 0.97 | 1.38 |
| 0.56 | 0.18 | 0.55 | 0.95 | 1.29 |
| 0.73 | 0.21 | 0.81 | 1.06 | 1.36 |
| 0.99 | 0.26 | 0.83 | 1.53 | 1.27 |

**Table II**

**Graphite Content of Nickel-Bearing Iron in Relation to Time at 900 Degrees Cent. (1650 Degrees Fahr.)**

| Per Cent Nickel | Per Cent Graphite Formed in Indicated Number of Hours at 900 Degrees Cent. (1650 Degrees Fahr.) | | | |
|---|---|---|---|---|
| | 2 Hours | 4 Hours | 6 Hours | 12 Hours |
| none | 0.06 | 0.22 | 0.70 | 1.23 |
| 0.24 | 0.08 | 0.42 | 0.97 | 1.11 |
| 0 46 | 0.18 | 0.46 | 1.03 | 1.25 |
| 0.66 | 0.13 | 0.59 | 1.05 | 1.24 |
| 0.91 | 0.17 | 0.66 | 1.06 | 1.29 |

**Table III**

**Graphite Content of Copper- and Nickel-Bearing Irons After 220 Hours at 700 Degrees Cent. (1290 Degrees Fahr.)**

| Per Cent Copper | Per Cent Graphite | Per Cent Nickel | Per Cent Graphite |
|---|---|---|---|
| trace | 0.46 | none | 0.46 |
| 0.33 | 1.04 | 0.24 | 0.83 |
| 0.56 | 0.79 | 0.46 | 0.84 |
| 0.73 | 0.77 | 0.66 | 0.90 |
| 0.99 | 1.01 | 0.91 | 1.09 |

**Table IV**

**Size Distribution of Nodules of Temper Carbon Formed at 900 Degrees Cent. in Copper-Bearing Iron**

| Per Cent Copper | Dia. Max. Nodule mm. | Nodules Per Cubic mm. | | | | | | Packing Density |
|---|---|---|---|---|---|---|---|---|
| | | Total | Max. Dia. | 4/5 Max. | 3/5 Max. | 2/5 Max. | 1/5 Max. | |
| trace | 0.17 | 679 | 40 | 62 | 95 | 111 | 371 | 14 |
| 0.33 | 0.11 | 1214 | 59 | 71 | 179 | 140 | 765 | 36 |
| 0.56 | 0.14 | 1319 | 47 | 73 | 59 | 55 | 585 | 24 |
| 0.73 | 0.14 | 787 | 40 | 63 | 182 | 105 | 377 | 22 |
| 0.99 | 0.14 | 1057 | 55 | 74 | 61 | 84 | 783 | 32 |

**Table V**
**Size Distribution of Nodules of Temper Carbon Formed at 900 Degrees Cent. in Nickel-Bearing Iron**

| Per Cent Nickel | Dia. Max. Nodule mm. | Nodules Per Cubic mm. Total | Max. Dia. | 4/5 Max. | 3/5 Max. | 2/5 Max. | 1/5 Max. | Packing Density |
|---|---|---|---|---|---|---|---|---|
| none | 0.17 | 679 | 40 | 62 | 95 | 111 | 371 | 14 |
| 0.24 | 0.15 | 784 | 57 | 53 | 96 | 51 | 526 | 15 |
| 0.46 | 0.17 | 1358 | 36 | 54 | 109 | 191 | 968 | 14 |
| 0.66 | 0.17 | 9297 | 44 | 101 | 117 | 176 | 8860 | 6 |
| 0.91 | 0.17 | 1527 | 20 | 83 | 130 | 173 | 1120 | 13 |

**Table VI**
**Size Distribution of Nodules of Temper Carbon Formed at 700 Degrees Cent. in Copper-Bearing Iron**

| Per Cent Copper | Dia. Max. Nodule mm. | Nodules Per Cubic mm. Total | Max. dia. | 4/5 Max. | 3/5 Max. | 2/5 Max. | 1/5 Max. | Density of Packing Per Cent |
|---|---|---|---|---|---|---|---|---|
| trace | 0.14 | 229 | 15 | 40 | 55 | 95 | 24 | 17 |
| 0.33 | 0.17 | 166 | 24 | 39 | — 4 | 19 | 88 | 26 |
| 0.56 | 0.17 | 44 | 24 | 8 | 15 | 4 | — 7 | 28 |
| 0.73 | 0.16 | 99 | 3 | 23 | 6 | 33 | 34 | 61 |
| 0.99 | 0.16 | 92 | 33 | 10 | 19 | — 3 | 33 | 35 |

**Table VII**
**Size Distribution of Nodules of Temper Carbon Formed at 700 Degrees Cent. in Nickel-Bearing Iron**

| Per Cent Nickel | Dia. Max. Nodule mm. | Nodules Per Cubic mm. Total | Max. dia. | 4/5 Max. | 3/5 Max. | 2/5 Max. | 1/5 Max. | Density of Packing Per Cent |
|---|---|---|---|---|---|---|---|---|
| None | 0.14 | 229 | 15 | 40 | 55 | 95 | 24 | 17 |
| 0.24 | 0.15 | 398 | 16 | 65 | 111 | 25 | 182 | 27 |
| 0.46 | 0.15 | 145 | 44 | 43 | 20 | — 2 | 40 | 18 |
| 0.66 | 0.15 | 490 | 71 | 13 | 0 | 60 | 346 | 17 |
| 0.91 | 0.13 | 510 | 109 | 73 | 48 | 35 | 241 | 17 |

of the 2-hour time, especially in the nickel series, is poor, suggesting a different and higher coefficient but not a fractional one as reported in (1). The lines of Fig. 1 were put in giving, as indicated, most weight to the higher graphite amounts. An exception is that for the alloy-free iron in which consideration was given to the data plotted in Fig. 2. In that figure the 1-hour intercepts of the lines of Fig. 1 are plotted against alloy content and are found to fall well along a straight line for each alloy. The intercept of these two lines with zero alloy content was considered in plotting the zero alloy line of Fig. 1. The concordance of these data gives us greater confidence in the interpretation than we could otherwise have.

Lacking data at several times we can evaluate the coefficient of graphitizing rate at 700 degrees Cent. (1290 degrees Fahr.) only by considering it proportional to the graphite formed in 220 hours. Here the copper data scatter erratically, the nickel data being somewhat more consistent. This is somewhat the opposite of the state of affairs at 900 degrees Cent. (1650 degrees Fahr.).

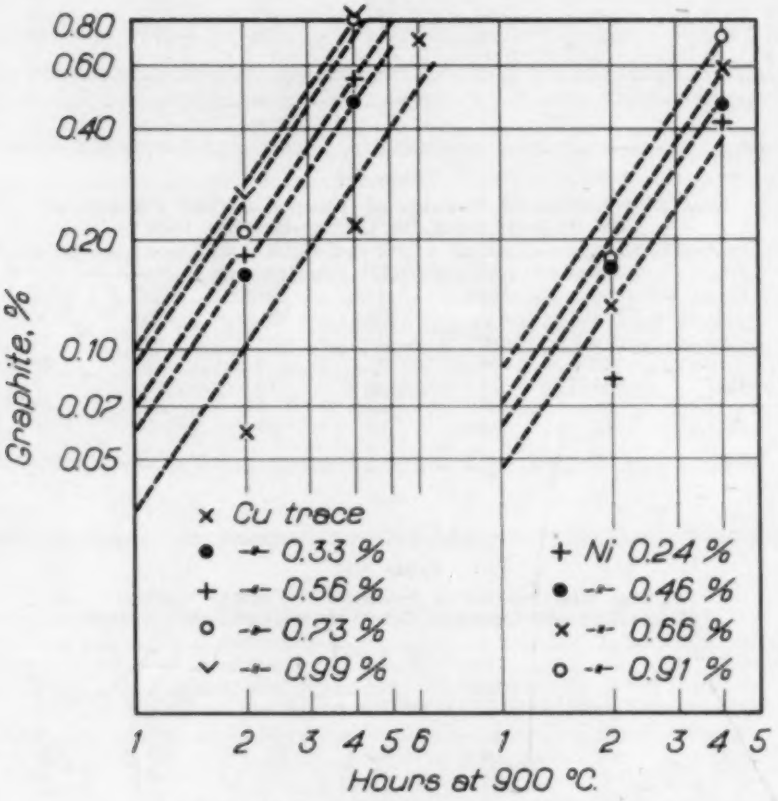


Fig. 1—Graphic Evaluation of Graphite-Time Data.

The data for 700 degrees Cent. (1290 degrees Fahr.) are plotted in Fig. 3 and scatter rather badly. Possibly they may be best interpreted by discarding the value for 0.33 copper as being affected by some unknown cause of error and representing the data for both copper and nickel by a single line here taken as straight because we have not sufficient reason to assign a definite, more complex form.

Taking copper and nickel as part of the same system is somewhat justified by the array of points in Fig. 2. Practically the sole alternative to this assumption would be to regard the first addition of alloy as highly effective and the subsequent addition as much less so. Be it noted that making the simpler assumption of a rectilinear relation of alloy content and coefficient of graphitizing rate 1 per cent of

either alloy multiplies the graphitizing rate by 110/46 = 2.2 — at 700 degrees Cent. (1290 degrees Fahr.) and by 100/35 = 2.9 — at 900 degrees Cent. (1650 degrees Fahr.). Further investigation on the effect of these alloys on graphitizing rate below $A_1$ would be desirable.

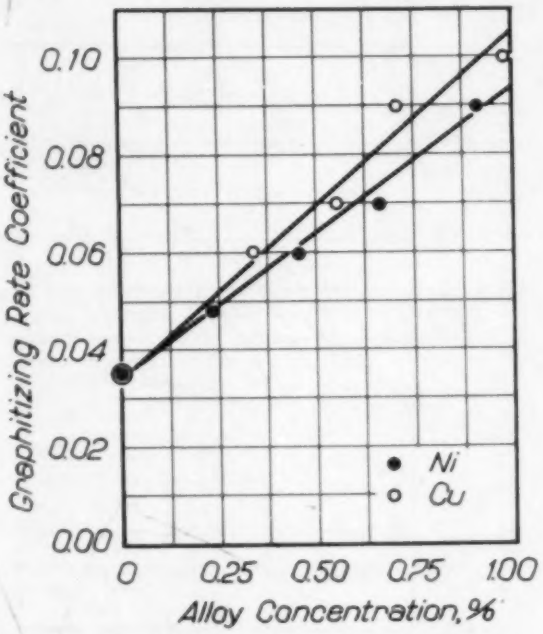


Fig. 2—Coefficient of Graphitizing Rate at 900 Degrees Cent. as a Function of Alloy Content.

Scrutiny of Tables IV and V discloses that at 900 degrees Cent. (1650 degrees Fahr.) the maximum nodule diameter is substantially the same, irrespective of alloy content. Also that in the copper alloys, except the highest, there is a somewhat uniform increase in nodule number with increasing alloy content. As is frequently the case, this number is determined largely by the number of nodules of smallest size which contribute but little to the graphite content. This condition, not at all unusual, is nevertheless unsatisfactory because there is always doubt as to whether small particles are actually graphite. Remembering that other things being equal, graphitizing rate should be proportional to nodule number, Tables VIII and IX have been compiled dividing the coefficient of graphitizing rate by nodule number and by the number of the four largest sizes of nodules.

Despite much uncertainty of interpretation due to the scatter of results to be expected in such investigations as these, the writer interprets these tables to mean that both alloys increase the graphitizing rate per nodule, i.e., the migratory rate of carbon in the metal.

Similar tables can be computed for 700 degrees Cent. (1290 degrees Fahr.).

One is inclined to interpret Table X as representing a decrease in nodule number and an increase in migratory rate with increasing copper content and Table XI as an increase in nodule number with-

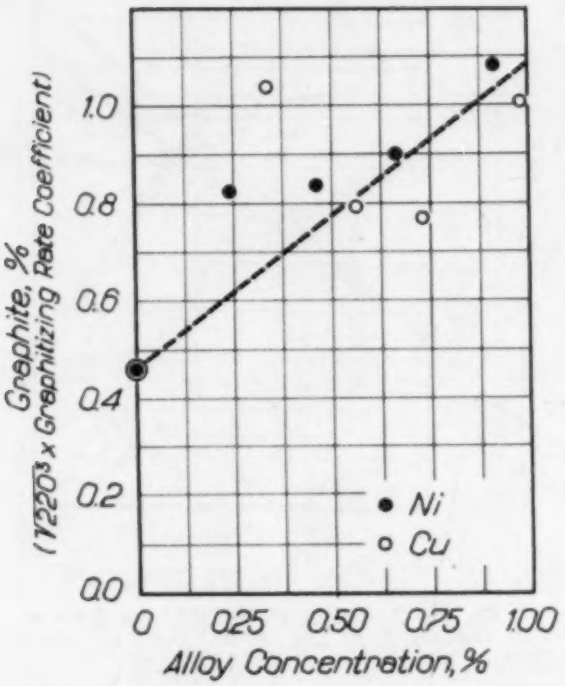


Fig. 3—Coefficient of Graphitizing Rate at 700 Degrees Cent. as a Function of Alloy Content.

out much change in migratory rate with increasing nickel content. In view of the irregularity of results this conclusion must however be only tentative. The reader should recall that in most previous work the inconsistencies are greater in observations below $A_1$ than above $A_1$. The conclusions regarding nickel are better supported than those regarding copper in this temperature range. No minimum graphitizing rate for an intermediate nickel content (0.25 per cent in the reference) was here observed.

Since the graphite content after 12 hours in Tables I and II are very similar for all alloy concentrations, it seems a fair assumption that the stable solubility of carbon in gamma iron is not materially altered by nickel or copper in the amounts here investigated.

The low density of packing of graphite at 700 degrees Cent. (1290 degrees Fahr.), which is the rule rather than the exception in

**Table VIII**

**Nodule Number and Coefficient of Graphitizing Rate at 900 Degrees Cent. (1650 Degrees Fahr.) for Copper-Bearing Iron**

| Copper Content | Nodule Number Per Cubic mm. Total | 4 Largest Sizes | Graphitizing Coefficient | $10^{\times}$ Graphitizing Coefficient Divided by Total No. | No. of Largest Sizes |
|---|---|---|---|---|---|
| 0.00 | 679 | 308 | 0.035 | 51 | 114 |
| 0.33 | 1214 | 449 | 0.060 | 49 | 134 |
| 0.56 | 1319 | 734 | 0.070 | 43 | 95 |
| 0.73 | 787 | 410 | 0.090 | 114 | 220 |
| 0.99 | 1057 | 274 | 0.100 | 95 | 365 |

**Table IX**

**Nodule Number and Coefficient of Graphitizing Rate at 900 Degrees Cent. (1650 Degrees Fahr.) for Nickel-Bearing Iron**

| Nickel Content | Nodule Number Per Cubic mm. Total | 4 Largest Sizes | Graphitizing Coefficient | $10^{\times}$ Graphitizing Coefficient Divided by Total No. | No. of Largest Sizes |
|---|---|---|---|---|---|
| 0.00 | 679 | 308 | 0.035 | 51 | 114 |
| 0.24 | 784 | 258 | 0.046 | 59 | 178 |
| 0.46 | 1358 | 390 | 0.060 | 44 | 158 |
| 0.66 | 9297 | 437 | 0.070 | 8 | 160 |
| 0.91 | 1527 | 407 | 0.090 | 59 | 221 |

**Table X**

**Nodule Number and Coefficient of Graphitizing Rate at 700 Degrees Cent. (1290 Degrees Fahr.) for Copper-Bearing Iron**

| Per Cent Copper | Nodule No. Per Cubic mm. Total | 4 Largest Sizes | Graphite in 200 Hrs. | $\sqrt{220^3} \times 10^4 \times$ Graph. Coefficient Divided by Total No. | No. of 4 Largest Sizes |
|---|---|---|---|---|---|
| 0.00 | 229 | 205 | 0.46 | 20 | 22 |
| 0.33 | 166 | 78 | 1.04 | 63 | 134 |
| 0.56 | 44 | 51 | 0.79 | 180 | 155 |
| 0.73 | 99 | 65 | 0.77 | 78 | 118 |
| 0.99 | 92 | 59 | 1.01 | 110 | 171 |

**Table XI**

**Nodule Number and Coefficient of Graphitizing Rate at 700 Degrees Cent. (1290 Degrees Fahr.) for Nickel-Bearing Iron**

| Per Cent Nickel | Nodule No. Per Cubic mm. Total | 4 Largest Sizes | Graphite in 200 Hrs. | $\sqrt{220^3} \times 10^4 \times$ Graph. Coefficient Divided by Total No. | No. of 4 Largest Sizes |
|---|---|---|---|---|---|
| 0.00 | 229 | 205 | 0.46 | 20 | 22 |
| 0.24 | 398 | 373 | 0.83 | 21 | 22 |
| 0.46 | 145 | 147 | 0.84 | 58 | 57 |
| 0.66 | 490 | 430 | 0.90 | 18 | 21 |
| 0.91 | 510 | 465 | 1.09 | 21 | 23 |

Tables VI and VII, has not previously been observed in this laboratory. We do not wish to make any particular point of the fact, but casual inspection of the specimens under the microscope suggests that the nodules formed at 700 degrees Cent. (1290 degrees Fahr.) are actually denser than those formed at 900 degrees Cent. (1650 degrees Fahr.) which suggests that the apparently low density of packing is actually a low order of symmetry, the nodules at 700 degrees Cent. (1290 degrees Fahr.) being rather far from equiaxed. Again this information is offered by way of comment rather than as a well authenticated observation.

## Conclusions

The following conclusions apply only to copper and nickel added in the ladle in amounts up to 1 per cent to fuel-melted white cast iron of approximately the composition shown under experimental procedure. They may be further limited by other as yet unsuspected variables connected with the history of the metal. Within the scope of the present experiments only:

(a) The accelerating effect of copper and nickel on graphitization has been confirmed.

(b) The two elements are of very similar potency, weight for weight. Their effect in terms of weight per cent is obviously more nearly alike than in terms of atomic per cent.

(c) The effect of 1 per cent of either element is about equal to that of about 0.9 per cent silicon.

(d) Neither element materially alters the solubility of graphite in gamma iron.

(e) Copper has but little effect on the number of nodules per unit volume formed at 900 degrees Cent. (1650 degrees Fahr.).

(f) Nickel increases the number of very small nodules per unit of volume formed at 900 degrees Cent. (1650 degrees Fahr.) but is of little effect on the number of the larger sizes.

(g) Either element increases the migratory rate of carbon in gamma iron solution.

(h) It is probable, but not established, that at 700 degrees Cent. (1290 degrees Fahr.) copper decreases the nodule number and increases the migratory rate of carbon (in alpha iron) in a manner to produce a net acceleration of graphitization while nickel increases the nodule number without considerable change in migratory rate.

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3. H. A. Schwartz, H. H. Johnson and C. H. Junge; "The Retarding Effect of Certain Metallic Elements on Graphitization," TRANSACTIONS, American Society for Metals, Vol. 25, 1937, p. 609.
4. H. A. Schwartz; "Graphitization at Constant Temperature," TRANSACTIONS, American Society for Steel Treating, Vol. 9, 1926, p. 883.
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## DISCUSSION

**Written Discussion:** By D. P. Forbes, president, Gunite Foundries Corp., Rockford, Ill.

It is a great pleasure to read the paper by Dr. Schwartz and his co-authors on the interesting quantitative effect of copper and nickel on graphitization rates of white iron.

In spite of the fact that the tabulated results do not follow as perfect a pattern as one might wish in an investigation of this kind, I believe there is no question but that Dr. Schwartz interprets correctly the general effect of these two alloying elements.

The paper points out that the conclusions are limited "to copper and nickel added in the ladle in amounts up to 1 per cent to fuel-melted white cast iron of approximately the composition shown, etc." This limitation is very significant in that it shows that the experimenter realizes there is possibly a profound difference in the effect of alloying additions added in the ladle and those incorporated in the original charge, also that white cast iron melted with fuel may be different from white cast iron of the same analysis melted electrically. It will be interesting, indeed, when these aspects of white and malleable iron are thoroughly investigated.

In this paper the degree of graphitization is measured in per cent at the various intervals of time. The commercial operation of graphitizing is conducted in two stages—a high temperature stage corresponding roughly to 900 degrees Cent. (1650 degrees Fahr.), and a second stage consisting either of a slow cool through the critical temperature or a hold at or near the critical temperature for a period of hours.

Commercially, the first stage of graphitization at the high temperature need be carried on only long enough to completely decompose the massive cementite by solution in the austenitic matrix. It would be interesting if the figures and data in this paper were assembled on the basis of time for complete decomposition of massive cementite. This would indicate the number of hours which could be cut off the annealing cycle by the use of copper or nickel.

**Written Discussion:** By C. H. Lorig, metallurgist, Battelle Memorial Institute, Columbus, Ohio.

The accelerating effect of copper and nickel on graphitization of white iron has again been shown by the work described in this paper. The data were secured on white irons melted in a Brackelsburg furnace of normal, blackheart malleable analysis and annealed in an electric furnace which brought the charge to temperature in approximately 1.5 hours. In mentioning the origin of their materials and the time for heating to the annealing temperature, the authors have given consideration to the fact that both the method of melting and the time of heating to the annealing temperature can have an important effect on the rate of graphitizing of white iron. While the influence of these factors is now recognized, the principles underlying the manner in which the factors affect the graphitizing rate are still obscure.

In graphitizing studies conducted by Smith and Palmer[3] on several series of copper-bearing, white cast irons the marked effects of the melting practice and the time of heating to the annealing temperature as well as the effect of copper on the rate of graphitization of white cast iron were shown. Their studies include both dilatometric and metallographic investigations for detecting the progress of graphitization and were conducted on commercial, air-

**Time for Graphitization of White Irons at 925 Degrees Cent. (1700 Degrees Fahr.)**

| Iron No. | Type of Iron | Speed of Heating | Graphitizing Times at 925 Degrees Cent. for Completion—Per Cent 50 Per Cent Hrs. | Min. | 95 Per Cent Hrs. | Min. | 100 Per Cent Hrs. | Min. |
|---|---|---|---|---|---|---|---|---|
| | | *Dilatometric Investigation* | | | | | | |
| 1 | Commercial Iron (1.01% Si, 0% Cu) | Rapid | 5 | 0 | 10 | 30 | 14 | 0 |
| | | Slow | 2 | 10 | 5 | 20 | 7 | 10 |
| 2 | Commercial Iron (1.01% Si, 1.33% Cu) | Rapid | 1 | 40 | 4 | 20 | 6 | 20 |
| | | Slow | 1 | 0 | 2 | 50 | 4 | 10 |
| 3 | Electric Iron (0.92% Si, 0% Cu) | Slow | 4 | 10 | 7 | 0 | 10 | 30 |
| | | *Metallographic Investigations* | | | | | | |
| 4 | Electric Iron (1.27% Si, 0% Cu) | Rapid | .. | .. | .. | .. | 16+[a] | .. |
| 5 | Electric Iron (1.22% Si, 1.0% Cu) | Rapid | .. | .. | .. | .. | 6 | 30 |
| 6 | Electric Iron (1.29% Si, 1.42% Cu) | Rapid | .. | .. | .. | .. | 4 | 30 |

(a) Small amount of primary cementite still retained.

*Compositions of White Irons*

| No. | Per Cent Carbon | Per Cent Silicon | Per Cent Manganese | Per Cent Sulphur | Per Cent Phosphorus | Per Cent Copper |
|---|---|---|---|---|---|---|
| 1 | 2.40 | 1.01 | 0.31 | 0.075 | 0.144 | 0 |
| 2 | 2.40 | 1.01 | 0.31 | 0.075 | 0.144 | 1.33 |
| 3 | 2.78 | 0.92 | 0.27 | 0.024 | 0.127 | 0.07 |
| 4 | 2.43 | 1.27 | 0.31 | 0.028 | 0.141 | 0.042 |
| 5 | 2.31 | 1.22 | 0.25 | 0.023 | 0.110 | 1.00 |
| 6 | 2.36 | 1.29 | 0.26 | 0.022 | 0.123 | 1.42 |

Rapid heating to 925 degrees Cent. in 20 to 40 minutes.
Slow heating to 925 degrees Cent. in 2 hours, 10 minutes.
Time at 925 degrees Cent. does not include time to reach temperature.

[3]C. S. Smith, and E. W. Palmer, "Some Effects of Copper in Malleable Iron," American Institute of Mining and Metallurgical Engineers, Iron and Steel Division, Vol. 116, 1935, p. 363-385; Also some unpublished information.

furnace and experimental, electric furnace white cast irons. The air-furnace irons were made in a production shop with copper added in the ladle. The experimental irons were made in an indirect-arc, electric furnace from charges comprising mixtures of pig iron, washed metal, steel and copper. The extent to which the graphitizing rate may be modified by the speed of heating to the annealing temperature and the method of preparing the iron is illustrated by some of their data in the table on the preceding page.

It is immediately apparent that the rate of heating to the annealing temperature was all important. For the copper-free commercial iron this factor alone accounted for a two-fold difference in time at temperature for complete graphitization to occur. For the commercial iron with 1.33 per cent copper the difference was less, but still very noticeable. It is interesting to note that copper had a greater influence on accelerating the graphitizing rate for the irons rapidly heated than for the irons heated more slowly.

The electric furnace irons, for reasons still obscure, graphitized more slowly than commercial white irons. The electric furnace iron No. 4, for example, although having a higher silicon content graphitized less rapidly than the commercial, air-furnace iron No. 1, both irons being heated to the annealing temperature in the same time. Cases of unusual stability of the cementite in the first stage of graphitization have occurred with electric melted irons. This stabilizing influence of melting practice appears to be lessened by alloying the iron with copper.

In the conduct of graphitizing studies such as described in this paper and in the interpretation of results it is apparent that due regard must be given to a host of factors operating to alter the graphitizing rate. The authors, being aware of these factors, have made an excellent contribution to our knowledge of the graphitizing influences of two important accelerators.

### Authors' Reply

In connection with the discusser's re-emphasis of our limitation of our conclusions to metal made in a particular way, I would point out that the reference in the paper describes earlier work we did on nickel white irons made otherwise, and since the results are not alike, we feel reasonable assurance that the results of a nickel addition made in the manner of the present paper are different from the results of a nickel addition made to an electric furnace bath. We have in an earlier paper pointed out a similar difference with respect to the addition of manganese.

Mr. Forbes suggests the desirability of a different way of expressing our results in order that a judgment might be formed as to the time saved by the addition of nickel. He would like to have information as to the time when the destruction of cementite is complete which in our notation is when the reaction is 100 per cent complete. It has perhaps escaped the discusser's notice that 100 per cent completeness of the reaction represents the graphitization of all of the carbon above that which remains in gamma iron solution.

Now we have shown in a very early paper on "Graphitization at Constant Temperature", TRANSACTIONS of the American Society for Steel Treating, Vol. 10, page 53, (1926), that the time-graphite curves expressed in terms of this

criterion of completeness are very similar. The rates of graphitization can be measured by the constant which we determined graphically by the intersection of the time-graphite curve with the 1-hour time ordinate.

Graphitizing rates are proportional to these coefficients and graphitizing times inversely proportional to the coefficients. The time to attain complete graphitization for any other specified degree of completion is therefore proportional to the reciprocal of these constants. It is also true that by consulting the time-graphite curves of the reference a reader can extrapolate from the observed degree of completion of graphitization in a known time to the time required for any other degree of completeness.

We have never in this laboratory been greatly interested in direct observations of the time when cementite completely disappears because the rate of disappearance of cementite decreases as time goes on. Theoretically the last traces never disappear and the detection or estimation of residual traces of carbide becomes so uncertain that enormous errors in time could readily be made.

The measurement of graphitizing rate is most sensitive when approximately one-half of the carbon in excess of that soluble at the existing temperature is converted into graphite. We prefer to make measurements in that region rather than in a region of greater uncertainty.

It may not be inopportune to point out that during the current convention a paper by Ziegler and his associates dealing with the problem of graphitization is being offered. This writer has presented written discussion of that paper bearing on the form of the time-graphite curve. It may be that that discussion would be of interest in connection with Forbes' timely comments.

Dr. Lorig has added a considerable amount of valuable information dealing with an aspect of the problem not included in its original scope. He has given us the results of investigations dealing with heating rate as a variable in determining graphitizing rate. A good deal of work has been done in various organizations exploring this variable but, as Dr. Lorig suggests, there is not yet any very satisfactory explanation of the fundamental mechanism. A tentative explanation was put forward by C. H. Junge of this writer's laboratory in a discussion of a paper by A. L. Boegehold at the 1938 meeting of the A.F.A. It seems somewhat probable that the explanation should lie in the direction of a nucleation phenomenon but no complete study has yet been made available. Indeed it is true that the observations of different competent observers are not yet altogether in harmony as to the fundamental facts.

It may be well to point out again that in an earlier series of investigations on nickel in which the manner of the addition of that element and the type of iron were different, a consistent series of results was obtained not in harmony with those resulting in the present investigation. This writer regards this as positive proof that alloying elements affect graphitizing rates by more than one mechanism each and that therefore their net effect may be a highly complicated result of several causes.

We are indebted to Dr. Lorig for the considerable factual contribution he has made to the subject in his discussion.

# HARDENABILITY VARIATIONS IN ALLOY STEELS—SOME INVESTIGATIONS WITH THE END-QUENCH TEST

By Gordon T. Williams

*Abstract*

*In the first part of the paper, results of several series of tests show that hardenability revealed by the end-quench method of Jominy and Boegehold is non-critical as to length or diameter of specimen, as is to be expected, since the quench is nearly ideal; quenching water temperature may vary between 35 and 145 degrees Fahr. with but minor change in resultant hardening, and use of NaOH solution instead of water has no measurable effect. Hardness gradients due to air-hardening effects were not found on traverses of the slower cooled ends of test bars. Shallow hardening carbon steels develop such a narrow martensitic band that accurate discrimination is difficult, and it is suggested that the major application of this test is to deeper hardening steels of which hardness traverses on quenched rounds of moderate section do not reveal sufficient differences.*

*The second part of the paper demonstrates that steels closely similar in chemical analysis for principal elements may require widely differing times at heat before full available hardenability is attained, but when this is achieved, similar curves are obtained. S.A.E. 3140 steels have been found to vary in this requirement from less than 30 minutes at 1500 degrees Fahr. to more than 8 hours at 1700 degrees Fahr. Curves are presented showing response of individual steels to variation in temperature and time at heat before quenching, as well as effect of previous structure. It is suggested that homogenization of austenite by diffusion is necessary for maximum hardenability and that degree of banding may be associated with the time-temperature requirement. Curves showing response of five typical heats each are given for S.A.E. 3140, 4150, and 4340.*

*Occasionally steels have been found to manifest a "hump" in the hardenability curve well away from the quenched end near Rc 35 and investigations on two such*

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. The author, Gordon T. Williams, is metallurgist, testing and research laboratories, Deere and Co., Moline, Illinois. Manuscript received June 23, 1939.

*steels are reported, the peculiarity being apparently related to the change in structure from intermediate constituent ("pseudomartensite") to pearlite.*

HARDENABILITY and its determination have been the subject of increasing attention during recent years, the 1938 American Society for Metals Symposium on "Hardenability of Alloy Steels" having brought out much information and revealed clearly the great need of further data. The several methods of testing steels for hardenability were described and discussed at that time. Essentially, there are two useful methods:

1. That of determining the hardness developed across a section from a round bar uniformly quenched from all sides (1).[1]

2. Determining hardness developed along a bar quenched on one end only (2), (3).

The first method is applicable to carbon and similar shallow hardening steels, but offers difficulty when differences are to be shown between steels which harden essentially throughout moderate sections. The second method, developed and described by Jominy and Boegehold (2), (3), reveals, on one test bar, hardening produced by a range of cooling rates representing every condition from that at the surface of a small bar drastically quenched to that at the center of a round bar several inches in diameter quenched in oil. The simplicity of this end-quench test, as well as the great amount of information readily obtained, make it a very valuable tool for investigation. The end-quench test was developed for carburized steel, for which it is in fact the only useful hardenability test, but it has been found even more useful for alloy steels of the direct hardening types.

The procedure used by the author in the tests described in this paper, is briefly as follows:

A test specimen 1 inch in diameter by 2¾ inches long is heated as desired, then quenched on one end face only, under controlled conditions. The test bars used in the author's work are made with a tapped hole in one end, into which a stainless steel bolt is inserted for handling and support during quenching. The flat opposite end of the specimen is smooth-polished, and rests during heating on a small carbon block to minimize scaling. When long time or high temperature is used, the carbon base tends to oxidize and expose

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

the bar end, so a steel tube is then placed around the specimen and carbon block, graphite powder or iron chips filling the space between the tube and the test bar. It should be noted that oxidation or decarburization of the specimen has no effect on the test result, if the end to be quenched is protected and provided that subsequent hardness readings are taken below any decarburization. The simple

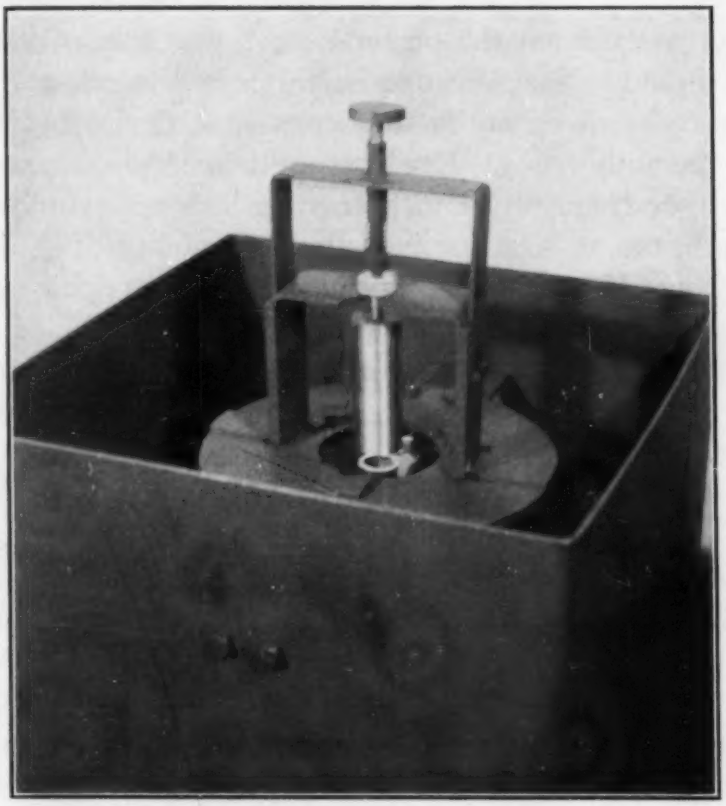

**Fig. 1—End-Quenching Apparatus.**

quenching equipment used by the author is shown in Fig. 1. It was easily and cheaply built, yet has repeatedly been found to give results identical with those from the more elaborate equipment which can be purchased. Above a small coolant-circulating pump is mounted a square tank; a 2-way valve on the side of the tank serves to direct the water from the pump, either through the ½-inch pipe against the specimen, or to a by-pass recirculating outlet. As may be seen in Fig. 1, a spring plunger holds the support-bolt head against a slotted cross bracket, thus accurately holding the specimen in fixed position yet permitting rapid placing. The quenched end is in the water stream, which spreads laterally with no tendency to wet the sides of the specimen. Height of water fountain, free, from the pipe, should be 2½

inches, according to Jominy and Boegehold (2), the specimen in place being ½ inch above the pipe opening, and water should be maintained at 75 degrees Fahr. As will be shown later, none of these details are critical, so those who built their own equipment need have no fear of inaccuracies affecting results to any important degree.

Rockwell Hardness tests are made on ground longitudinal flat surfaces; in many cases checks have been made on surfaces 90 or 180 degrees away from the original, no differences of consequence being thus found. The specimen is laid on a V-block and advanced under the Rockwell tester by the screw-slide device previously described by the author (4). Readings are taken $\frac{1}{16}$ inch apart for the first ½ inch, or frequently 1 inch, then ⅛ inch apart for the balance of the length, unless behavior indicates advisability of $\frac{1}{16}$ inch spacing throughout. The greatest change in cooling velocity occurs, of course, in the first few readings. The use of a hand crank on a 16-pitch screw greatly facilitates desired spacing of readings.

Since the prior microstructure is known to have a definite effect on hardenability (see also Fig. 7), the writer has adopted as standard procedure the preliminary annealing of all samples at 1600 degrees Fahr. (870 degrees Cent.) for 1 hour followed by furnace cooling; any conditioning treatment could be used, but this most closely approximates usual production practice. This anneal is given all samples, usually before machining, unless effect of special pre-structure is being investigated.

Hardness curves are drawn smoothly through points representing readings, variations of 1 point being ignored if on single readings. A smooth curve, it would appear, is more justifiable from our knowledge of what is occurring than a sharp, point-to-point graph, which is also not warranted by the spacing of readings.

It may be remarked here that it is unnecessary, in application of the test results, to determine cooling rates in specific locations in parts to be made from a given steel. An empirical solution is simply to run the end-quenched test on a sample of a certain steel, then to make hardness tests where desired on parts made from this same steel treated in production. The hardness attained in a selected spot results from the same cooling rate where that hardness is obtained on the test bar. All similar steels identically treated will develop, at the location in question, the hardness found at the established position on the test bar. Thus, $\frac{11}{16}$ inch along the bar may be found to show the same hardness as at some location in a quenched part on

one steel; therefore, these two locations cooled at the same rate, and any new lot of steel tested will show by its $\frac{1}{16}$ inch reading what will be found at the selected location in parts made from the new steel. In this relationship furnace and quenching conditions in production are added into the final result. Note, however, that it is not assured that wide variations of steel type can be covered from one such established relationship.

It may be mentioned that the end-quenched test bar lends itself admirably to tempering tests, the rate or degree of softening of all the zones of different hardness being readily followed on successively tested longitudinal axes.

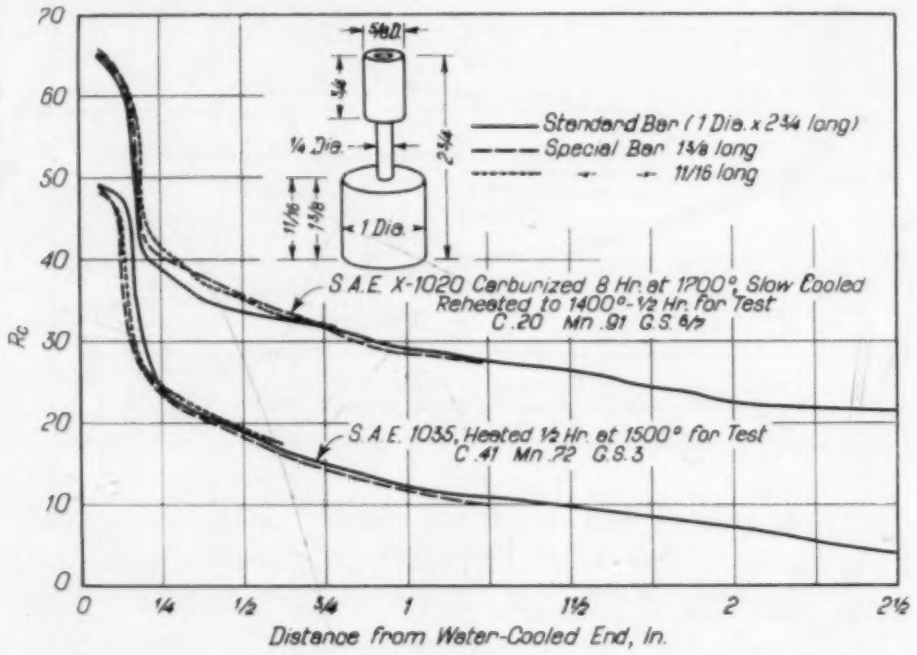
Standard Bar (1 Dia. x 2¾ long)
Special Bar 1⅝ long
11/16 long
S.A.E. X-1020 Carburized 8 Hr. at 1700°, Slow Cooled
Reheated to 1400°-½ Hr. for Test
C .20 Mn .91 G.S. 6/7
S.A.E. 1035, Heated ½ Hr. at 1500° for Test
C .41 Mn .72 G.S. 3
Rc
Distance from Water-Cooled End, In.

Fig. 2—Effect of Length of Test Bar on End-Quenched Hardenability.

## Part I. Influence of Various External Factors on Test Results

It was found in early use of this test method that shallow-hardening carbon steels gave a very narrow martensite band, and the undesirability of taking Rockwell readings closer together than $\frac{1}{16}$ inch makes accurate discrimination on such steels difficult. Consequently, a study of the basic factors of the method was undertaken to find out whether a greater depth of hardening could be obtained by some variation, and to disclose critical features.

Work of Scott (5), Grossmann (6) and others has shown that violently agitated water gives almost an ideal quench, which may be briefly described as cooling which brings the surface of the hot steel instantly to the coolant temperature and maintains it there, the limiting factor being the rate of heat transfer in the steel itself. Thus, briefly, the hardenability of a given steel in an ideal quench will be independent of thickness of metal if effect of radius of curvature is absent. A flat bar 1 inch thick or 12 inches thick would harden to the same depth and show the same hardness curve so far as it

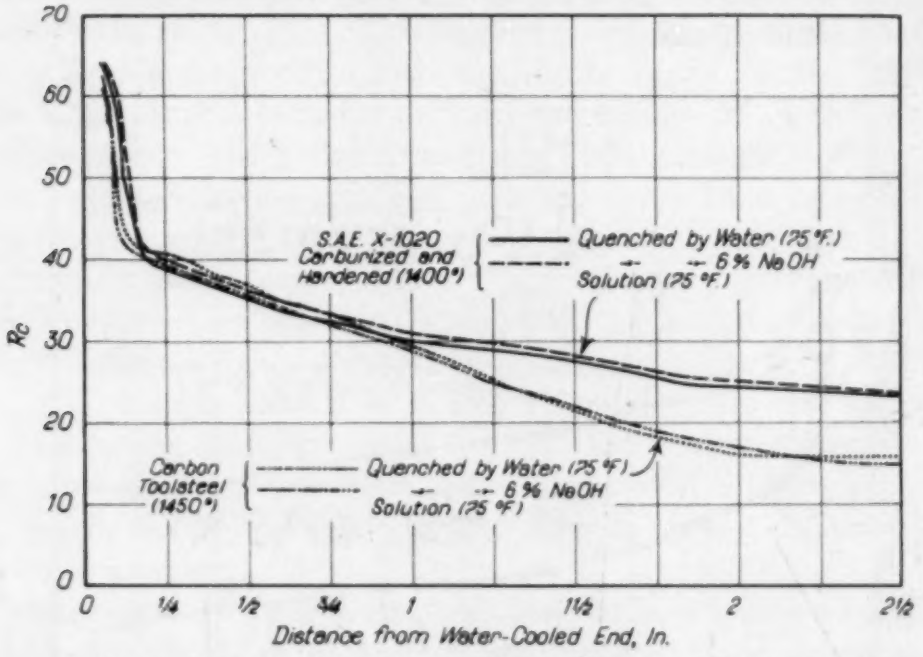


Fig. 3—Effect of Water Versus Caustic Soda Solution on End-Quenched Hardenability Standard Test Bar.

went, a flat having a curved surface of infinite radius. It may be expected that the quench in this hardenability test method is nearly ideal, and no real change can be effected by varying the quench or length of specimen: that this is true is clearly shown by the following tests. The complete absence of any effect due to change in length of the specimen is shown in Fig. 2. The two steels tested, S.A.E. 1035 and carburized S.A.E. X-1020, were made up into specimens having length of the 1 inch diameter portion of $\frac{1}{16}$ inch, 1⅜ inch and 2¾ inch; that is ¼, ½ and full standard in length, but the hardness curves were identical so far as each went. Similar results were attained on carbon tool steel. In Fig. 3 it is shown that substitution of 6 per cent sodium hydroxide solution had no effect on

hardenability of carbon tool steel or carburized S.A.E. X-1020. Fig. 4 shows results of varying quench-water temperature; 35, 75, 105 and 145 degrees Fahr. were all used in testing identical specimens of S.A.E. 3140. It will be noted that all except the 145-degree curve are close checks within the limits of accuracy of hardness testing perhaps, although even closer duplication is usually had when conditions are identical. The 145-degree curve is the lowest, but only

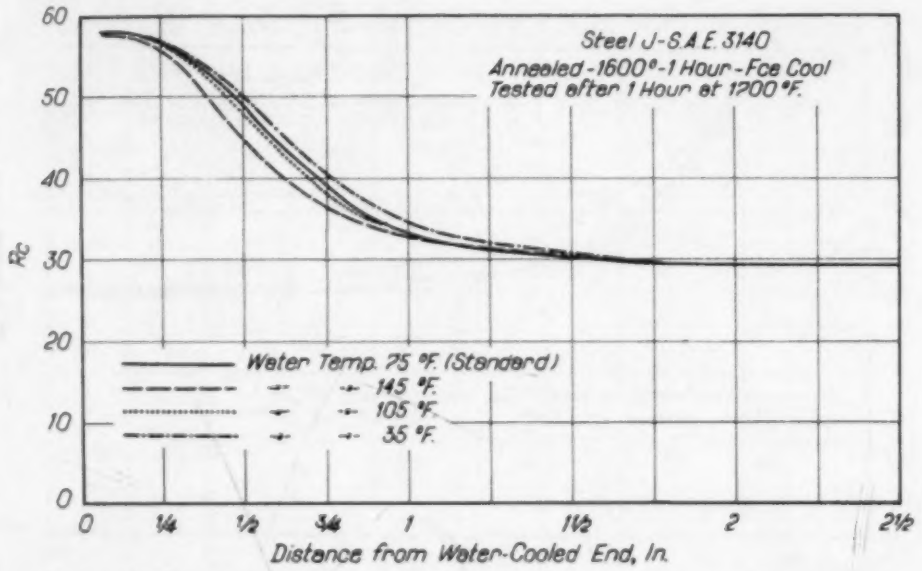


Fig. 4—Effect of Quench-Water Temperature on End-Quenched Hardenability.

in the transition zone between martensite and pearlite, and even here the greatest difference is 5 points Rc. This is a good demonstration of the reproducibility of results and the quality of the quench.

Variation in diameter of test specimens should be without essential effect, since end area and bar volume have constant ratio. The series shown by Fig. 5 demonstrated this clearly. S.A.E. 3140 was made into bars ½, 1, and 1½ inches in diameter, otherwise standard, and after end-quenching from identical heating conditions the curves were found to be practically coincident. The usual ½-inch water stream was used for both the 1 and 1½-inch bars, and the supply of quenching medium is obviously ample. The ½-inch bar showed slightly greater hardness through the transition zone; since the water outlet was restricted only to ¼ inch it is possible that some creeping of water up the sides of the specimen occurred. In any case the differences are negligible.

It was also considered advisable to determine whether there was

any hardness gradient from surface to center, due to an air cooling effect away from the quenched end. End-quenched specimens of several steels were cut transversely ½-inch and 1¼-inch away from the slow-cooled end and hardness traverses run in the ordinary manner. Fig. 6 shows typical plots of such tests on two steels. Clearly the effect of air cooling is inconsequential so far as formation of a hardness gradient is concerned. The air cooling rate is, of course, the lowest which can be found on any length bar.

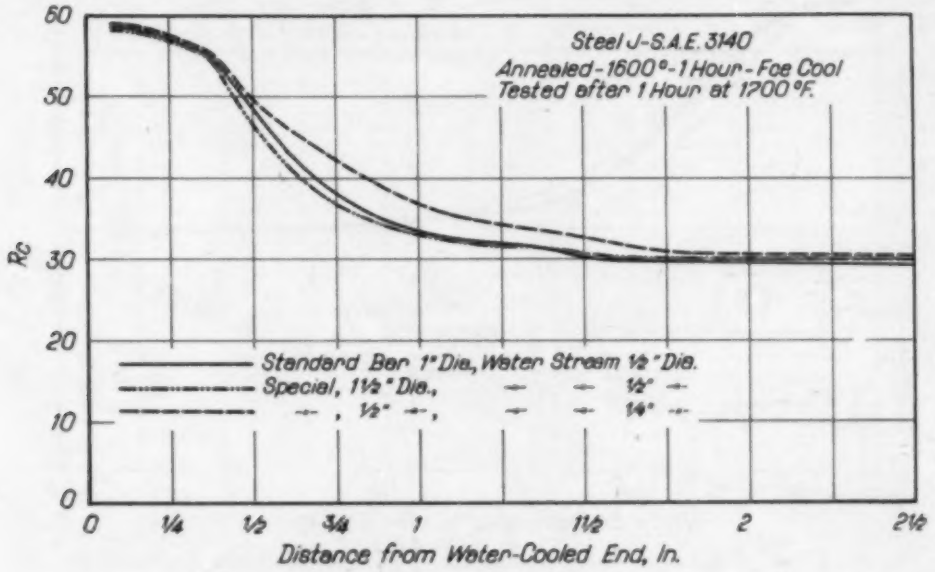


Fig. 5—Effect of Test-Bar Diameter on End-Quenched Hardenability.

From the above tests, it may be concluded that the end-quenched hardenability test is noncritical as to length and diameter of specimen or water temperature over a broad range; and that the effect of air cooling does not produce hardness gradients even at the slow-cooled end, so that depth below the surface is unimportant when taking hardness readings. Thus, if no decarburization is present readings from an extreme surface axis are satisfactory; and, likewise, the axis to be tested may be at any required or desired depth. Further, these tests point to no method whereby the depth of hard zone of carbon steels may be enlarged on the standard bar except the unsatisfactory process of cutting a hardened specimen at such an angle that the martensite zone is expanded. It may be remarked that when results are plotted with the logarithm of rate of cooling or distance along the bar as abscissa, a much more significant type of curve is obtained, but accuracy of discrimination between steels is

not improved, the curve being only mechanically enlarged at the higher hardness end.

## Part II. Hardenability of Typical Alloy Steels as Affected by Time and Temperature of Heating

The experiments discussed in Part I of this paper were undertaken only after repeated observation had shown that excellent reproducibility of hardenability curves was normally obtained. It was, however, early discovered that variations in time of holding at heat before end-quenching gave in some cases a wide variation in relative hardenability. To pursue this further, identical specimens were machined from one bar of cold drawn S.A.E. 3140, and pairs were sub-

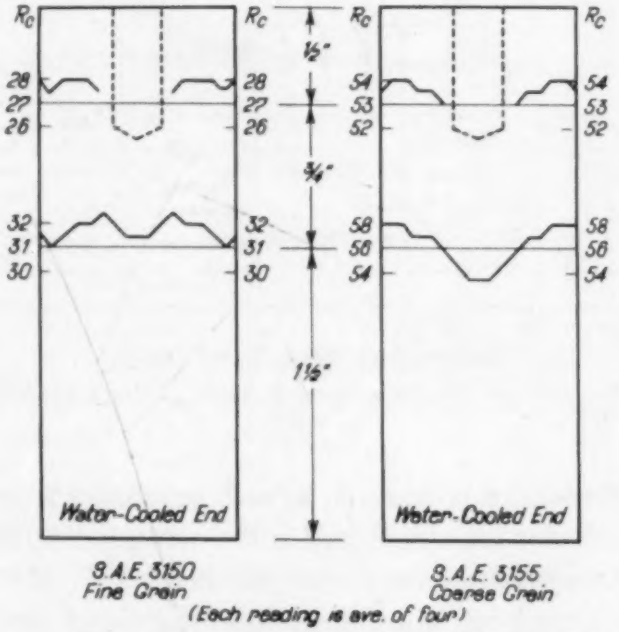


Fig. 6—Hardness Traverses of End-Quenched Hardenability Test Bars.

jected to four different pre-conditioning treatments, as follows: Normalized, quenched, quenched and tempered, and annealed (the detailed treatments are listed in Fig. 7). One pair was left as cold drawn. One piece in each condition was held at 1500 degrees Fahr. for ½ hour before end-quenching, the other held for 1½ hours. The resultant hardness curves are shown in Fig. 7. It is apparent that full hardenability was not attained on any of the specimens after the short time at heat, a marked increase being achieved by the 1½-hour soak. There is also given in Fig. 7 a clear indication of the effect of previous carbide particle size on degree of hardening at-

tained at shorter times. The lowest curve for ½-hour hold is obtained from the specimen previously coarsely pearlitic.

Since after the longer time at heat prior to testing all samples gave practically identical curves, it appeared likely that this type of curve, i.e., flat throughout the lower hardness range, might represent the "maximum"[2] hardenability attainable on a steel at any given temperature. Subsequent work has apparently confirmed that this is the

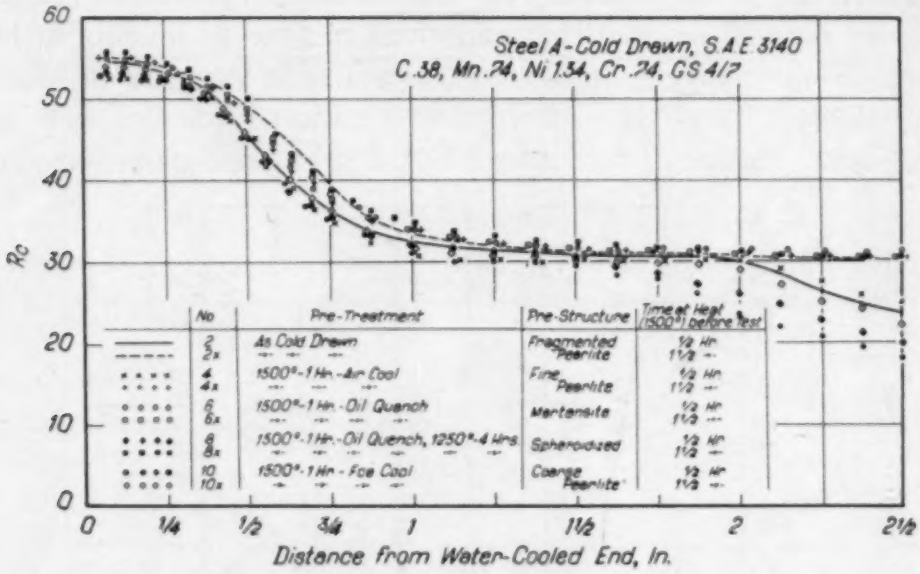


**Fig. 7—Effect of Pre-Structure and Time at Heat on End-Quenched Hardenability.**

case when carbide size is normal, as will be shown later. The effect of time at heat before testing is again shown in the series summarized in Fig. 8. Specimens all from one bar of S.A.E. 3140 were heated for 5, 30, 45, 90 and 180 minutes at 1500 degrees Fahr. (815 degrees Cent.) before testing and the hardness curves show a gradual but decreasing improvement. The 180-minute hardenability closely approaches the "maximum."

Microscopic examinations of the martensite zone of these and other specimens which developed less than the "maximum" hardenability found after longer times at heat were made in an effort to find carbides out of solution. In no case, unless the samples had been deliberately spheroidized for long periods, could carbide particles be seen. Harder and Lauderdale (7) have demonstrated that, in carbon

[2]The word "maximum" is here used in a special sense, i.e., the maximum hardenability which can be attained below the grain growth temperature, and is, therefore, enclosed in quotation marks. All steels herein reported in this respect are fine-grained by McQuaid-Ehn test, and 1700 degrees Fahr. was top hardening temperature used.

steels, carbide solution is extremely rapid, a matter of but seconds being required. They also showed that maximum hardness (in steel below 0.50 per cent carbon) is developed soon after carbides disappear. However, this work was confined to carbon steels in small sections in which no depth-hardening effects were to be found. Undoubtedly, the moderately alloyed steels have decreased carbide solubility velocity, but it may also be assumed that the time required

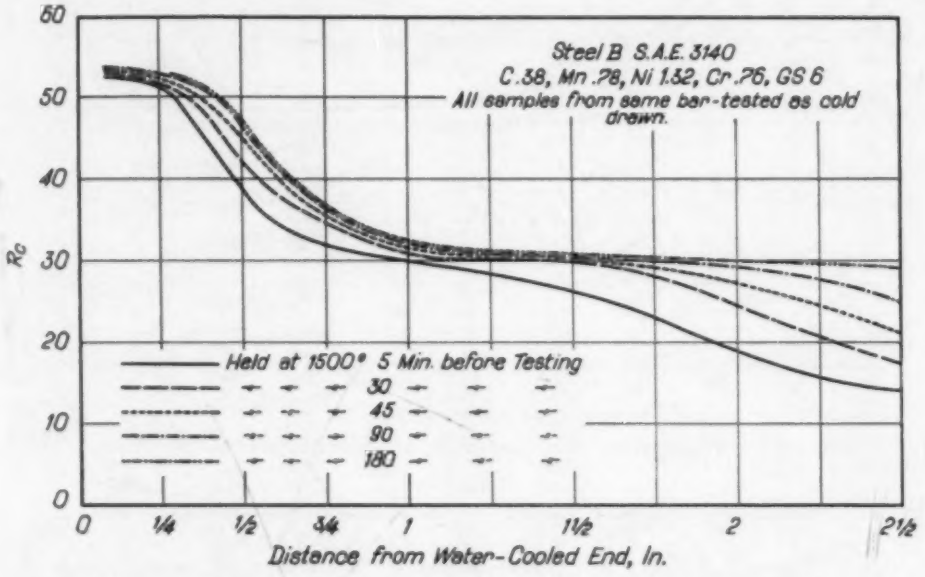


Fig. 8—Effect of Time at Heat on End-Quenched Hardenability.

is but a matter of a few minutes at normal heat treating temperatures for particles of usual size. Failure to find carbides out of solution in specimens used in the present investigation is thus in accord with the above mentioned work. In consequence of this, the author believes that the increasing hardenability with longer holding times at heat is a result of increasing homogenization of the austenite, to which there is obviously an upper limit.

Alloy steels essentially identical as regards normal chemical analysis (the influence of minor elements not usually determined may be vital) may show widely differing response to time at heat. Such action has been reported for high carbon-high chromium or high speed steels, but not for commercial alloy steels. Tests on samples from five representative commercial heats of S.A.E. 3140, all closely similar in analysis and grain size, have been selected to demonstrate this variation. In Fig. 9a, the hardenability curves for these steels are shown as end-quenched after ½ hour at 1500 degrees Fahr.

(815 degrees Cent.); a wide spread is revealed, more than to be expected from far greater differences in alloy content. That this spread of results after ½ hour at 1500 degrees Fahr. is entirely a function of holding time is shown by the curves in Fig. 9b, where the "maxi-

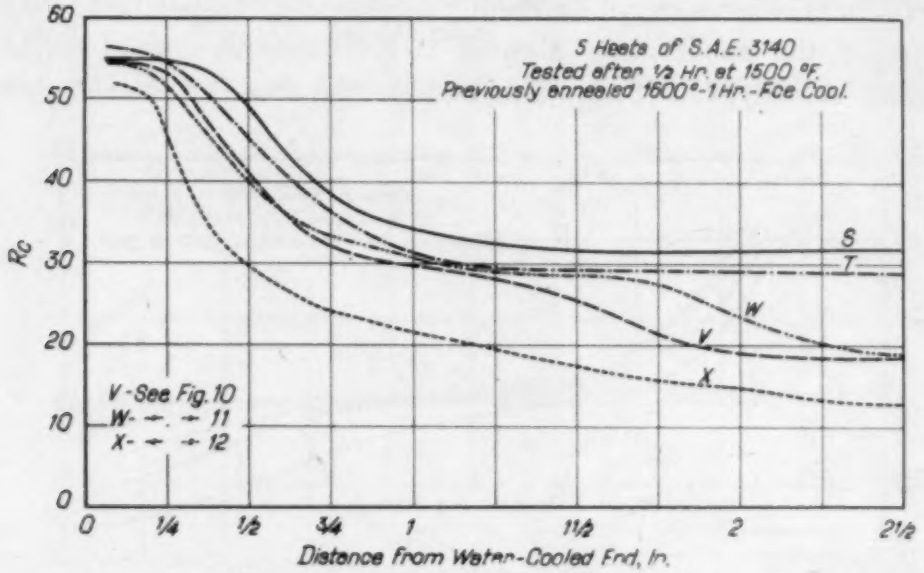


Fig. 9a—End-Quenched Hardenability of Five Heats of S.A.E. 3140 Steel.

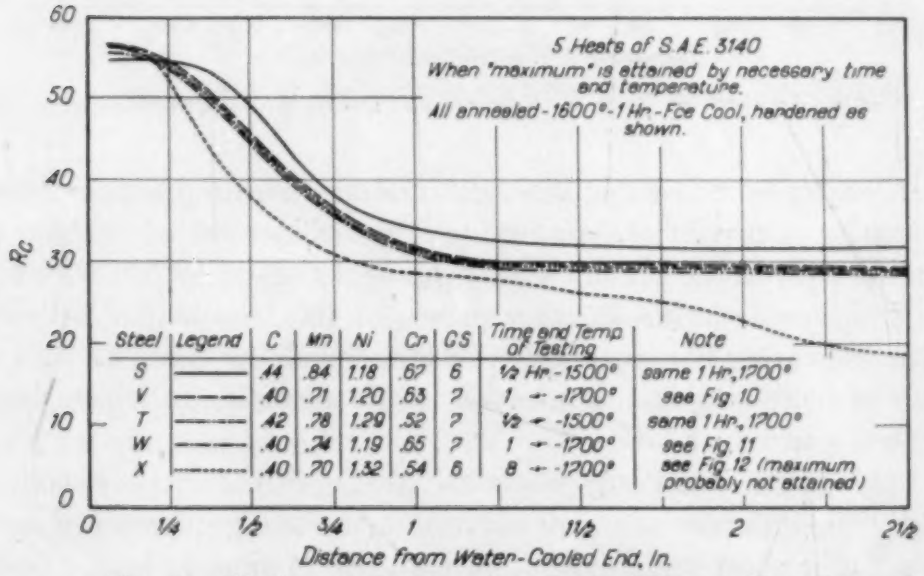
5 Heats of S.A.E. 3140
When "maximum" is attained by necessary time and temperature.
All annealed - 1600°-1 Hr.-Fce Cool, hardened as shown.
Rc
60
50
40
30
20
10
0
0
¼
½
¾
1
1½
2
2½
Distance from Water-Cooled End, In.
Steel | Legend | C | Mn | Ni | Cr | GS | Time and Temp. of Testing | Note
S | ——— | .44 | .84 | 1.18 | .62 | 6 | ½ Hr.-1500° | same 1 Hr., 1700°
V | — — — | .40 | .71 | 1.20 | .63 | 7 | 1 " -1700° | see Fig. 10
T | ——— | .42 | .78 | 1.29 | .52 | 7 | ½ " -1500° | same 1 Hr., 1700°
W | — — — | .40 | .74 | 1.19 | .65 | 7 | 1 " -1700° | see Fig. 11
X | ..... | .40 | .70 | 1.32 | .54 | 6 | 8 " -1700° | see Fig. 12 (maximum probably not attained)

Fig. 9b—"Maximum" End-Quenched Hardenability of Five Heats of S.A.E. 3140 Steel.

mum" hardenability developed by each steel is given. Two heats, S and T, have already attained their "maximum" hardenability after ½ hour at 1500 degrees Fahr., while V and W required 1 hour at 1700 degrees Fahr. (925 degrees Cent.) and X was still improving

after 8 hours at 1700 degrees Fahr., the "maximum" apparently not yet having been attained. In Figs. 10, 11, and 12 are given curves representing series of tests run on steels V, W, and X of Fig. 9. The rate of attainment of "maximum" hardenability is obviously widely different in these several steels. The normal analyses, listed in Table I, reveal no indication of cause, nor do McQuaid-Ehn grain sizes.

**Table I**
**Analyses of Steels Used in Part II**

| Spec. | Code | C | Mn | Si | P | S | Ni | Cr | Mo | N | McQ.-E. G. S. |
|---|---|---|---|---|---|---|---|---|---|---|---|
| S.A.E. 3140 | A | 0.38 | 0.74 | .... | ..... | ..... | 1.32 | 0.74 | .... | ..... | 4/7 |
| | B | 0.38 | 0.78 | .... | ..... | ..... | 1.32 | 0.76 | .... | ..... | 6 |
| | S | 0.44 | 0.84 | 0.21 | 0.019 | 0.032 | 1.18 | 0.67 | .... | 0.0053 | 6 |
| | T | 0.42 | 0.78 | .... | 0.026 | 0.025 | 1.29 | 0.52 | .... | ..... | 7 |
| | V | 0.40 | 0.71 | .... | 0.015 | 0.026 | 1.20 | 0.63 | .... | ..... | 7 |
| | W | 0.40 | 0.74 | .... | 0.021 | 0.028 | 1.19 | 0.65 | .... | ..... | 7 |
| | X | 0.40 | 0.70 | 0.22 | 0.017 | 0.025 | 1.32 | 0.54 | .... | 0.0067 | 6 |
| S.A.E. 3150 | J | 0.48 | 0.69 | .... | ..... | ..... | 1.39 | 0.56 | .... | ..... | 6 |
| | K | 0.47 | 0.77 | .... | ..... | ..... | 1.22 | 0.63 | .... | ..... | 6 |
| S.A.E. 4340 | L | 0.42 | 0.73 | 0.22 | 0.018 | 0.020 | 1.71 | 0.73 | 0.34 | 0.0053 | 6/7 |
| | M | 0.41 | 0.70 | .... | 0.016 | 0.019 | 1.77 | 0.71 | 0.35 | ..... | 7/6 |
| | N | 0.40 | 0.71 | .... | 0.017 | 0.021 | 1.76 | 0.68 | 0.34 | ..... | 6 |
| | O | 0.42 | 0.66 | .... | 0.018 | 0.015 | 1.78 | 0.67 | 0.34 | ..... | 7/6 |
| | P | 0.40 | 0.64 | 0.16 | 0.019 | 0.022 | 1.65 | 0.60 | 0.35 | 0.0081 | 6 |
| S.A.E. 4150 | C | 0.49 | 0.81 | 0.18 | 0.018 | 0.024 | .... | 0.97 | 0.24 | 0.0062 | 6 |
| | D | 0.51 | 0.86 | 0.18 | 0.018 | 0.025 | .... | 1.05 | 0.22 | 0.0062 | 6/7 |
| | E | 0.51 | 0.82 | .... | 0.020 | 0.031 | .... | 0.92 | 0.19 | ..... | 6 |
| | F | 0.48 | 0.84 | .... | 0.027 | 0.027 | .... | 1.03 | 0.20 | ..... | 6 |
| | G | 0.51 | 0.74 | .... | 0.022 | 0.016 | .... | 0.91 | 0.21 | ..... | 6/7 |
| Special | Q | 0.63 | 0.66 | .... | ..... | ..... | 1.15 | 0.58 | .... | ..... | 5 |
| | R | 0.55 | 1.19 | .... | 0.017 | 0.047 | 0.19 | 0.57 | 0.02 | ..... | 2 & 5 |

Microstructure of identically annealed samples of each of these steels is shown in Fig. 13. It is apparent that the degree of banding revealed in each steel is in line with its hardenability behavior. Such might be expected to be the case, banding being a manifestation of nonhomogeneity and of resistance to diffusion. It is reasonable to believe that the two features are closely related. Metallographic examination of the quenched specimens revealed a decrease in banding as "maximum" hardenability was approached.

The relative effectiveness of various temperatures in bringing about homogenization is indicated by the results of several series of tests on steels which showed very slow attainment of "maximum" hardenability. Partial homogenizations gave similar curves regardless of temperature, time being, of course, different. Ten hours at 1500 degrees Fahr., 2.5 hours at 1550 degrees Fahr., and ½ hour at 1600 degrees Fahr. were found to give identical curves on one steel; ¼ hour at 1650 degrees Fahr. was more effective than one hour at

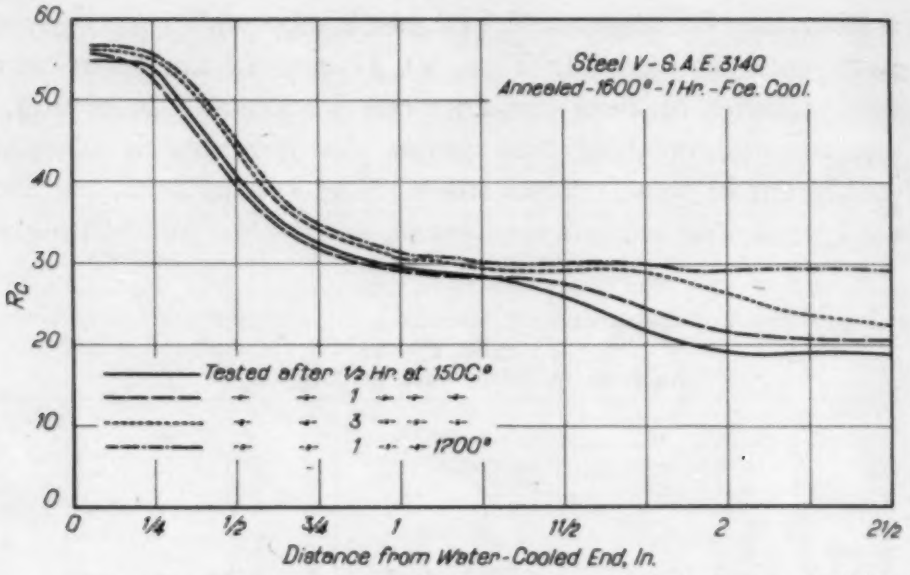


Fig. 10—Variation in Time-Temperature Requirements for Attaining "Maximum" End-Quenched Hardenability.

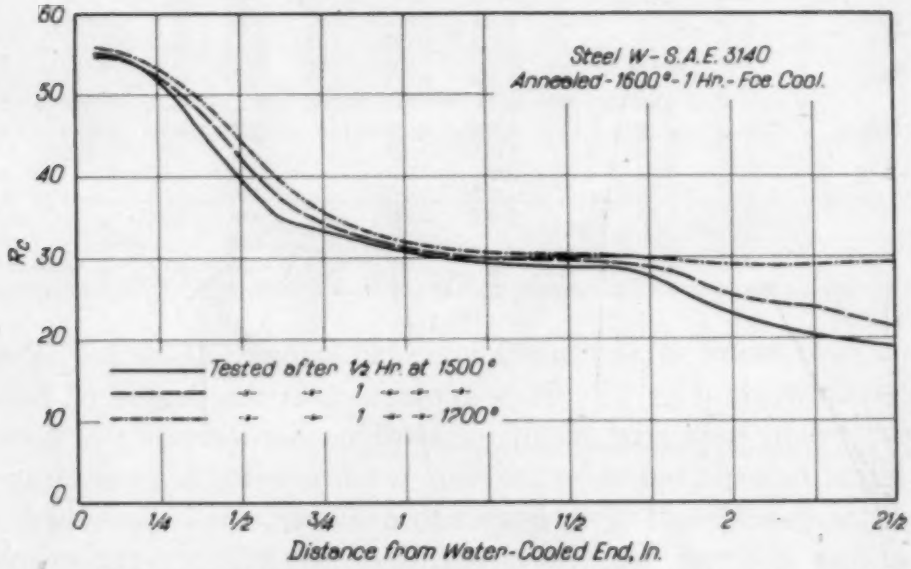


Fig. 11—Variation in Time-Temperature Requirements for Attaining "Maximum" End-Quenched Hardenability.

1600 degrees Fahr. These ratios are not unexpected, but point out clearly the tremendous differences in time requirements in the steels of Fig. 9. Eight hours at 1700 degrees Fahr. is undoubtedly hundreds of times as effective as ½ hour at 1500 degrees Fahr., yet more than this difference is shown between steels S or T and X.

Other types of alloy steels have also been found to show varying response to time and temperature. Fig. 14 shows the summary of two heats of S.A.E. 3150 of similar chemistry, but widely differing

response to time at heat. Figs. 15 and 16 give detailed series of curves for these two steels. It may be remarked that these two are steels whose transverse hardening behavior was reported in a previous paper (4). The response to tempering of two test specimens of the same steel, one incompletely homogenized, and the other heated long enough to attain "maximum" hardenability, is shown in Fig. 17.

The hardenability of five heats of S.A.E. 4340 is represented by the curves of Fig. 18. All of these steels are closely similar in

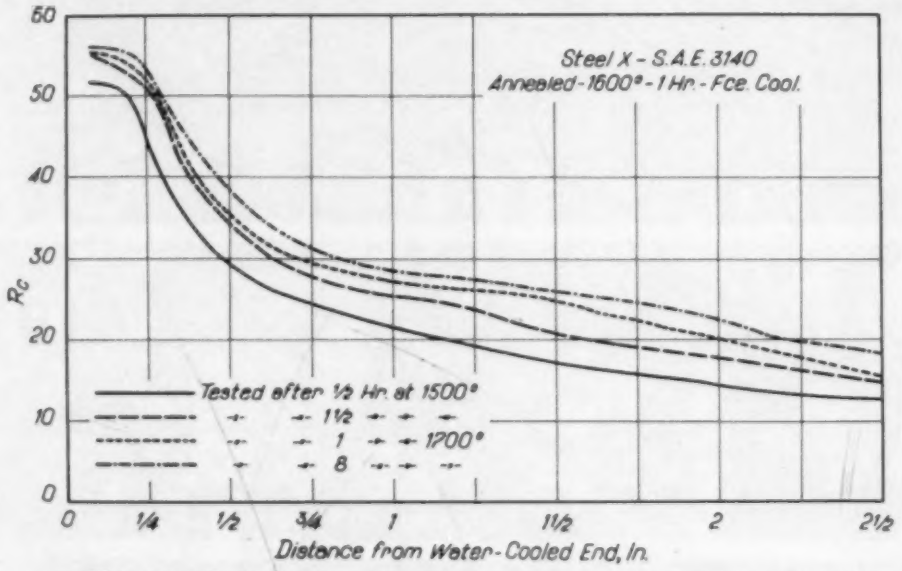


**Fig. 12—Variation in Time-Temperature Requirements for Attaining "Maximum" End-Quenched Hardenability.**

common alloys, as well as sulphur, phosphorus, nitrogen and silicon. One heat particularly shows surprisingly low hardenability. It is of interest to note that this heat was originally discovered by a production heat treating department to be low in as-quenched hardness, so that draw temperature had to be severely decreased to meet hardness specifications. In fact, water quenching was resorted to in order to raise the draw temperature to a desirable range. Shallow hardening, of course, connotes greater sensitivity to size and quenching conditions. This particular steel has been end-quenched from temperatures as high as 2100 degrees Fahr. (1150 degrees Cent.), where grain size was 2, without improvement in hardenability curve. Analysis, including S, P, N, Si, is normal. Obviously, some unknown factor is here present. A definite increase in hardness at the lower end of the curve was found when a specimen heated to 1550 degrees

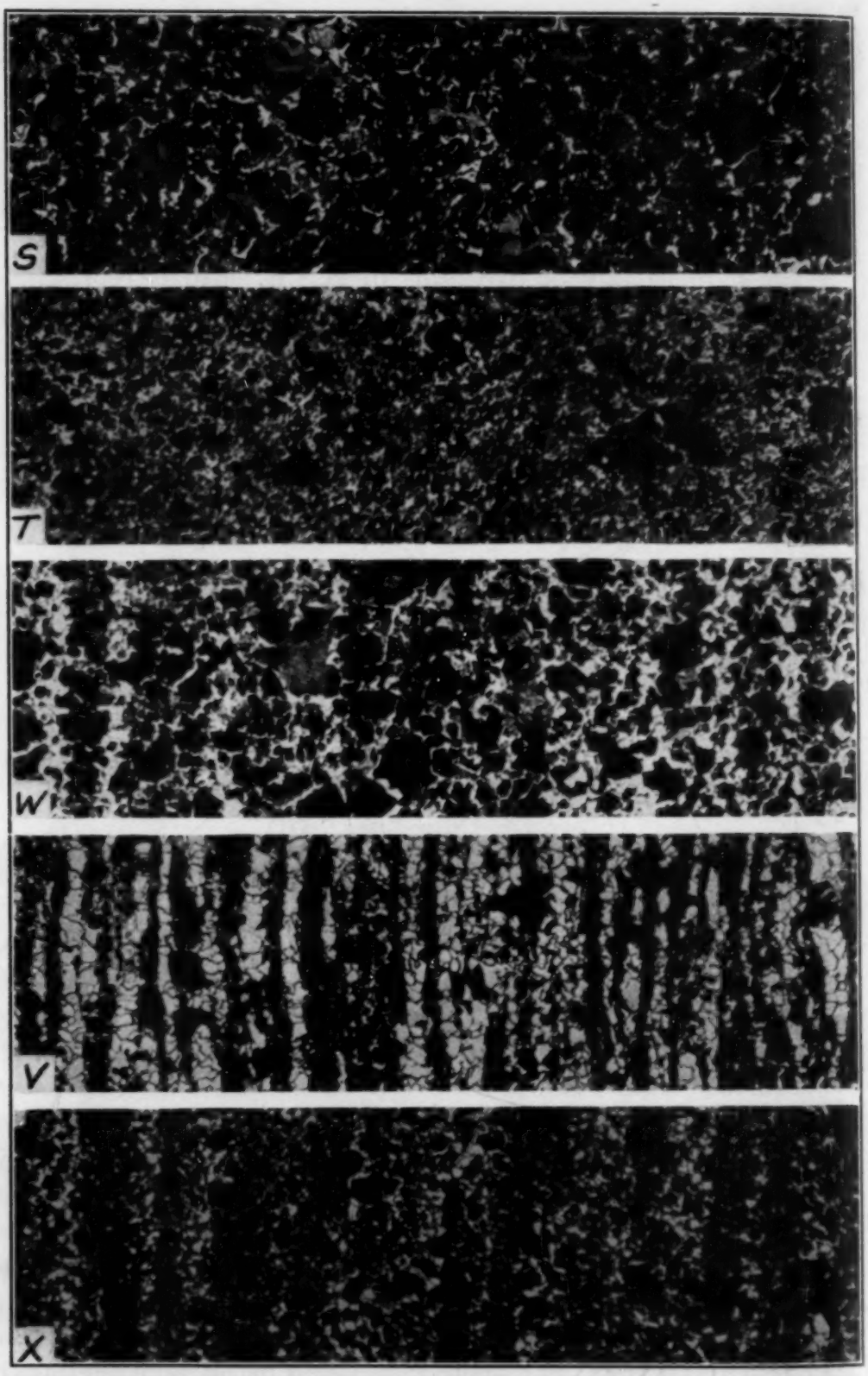


Fig. 13—S.A.E. 3140 Steels, Annealed 1 Hour at 1600 Degrees Fahr. Furnace Cooled. Time Required to Reach Maximum Hardenability Increasing Toward Bottom. × 100.

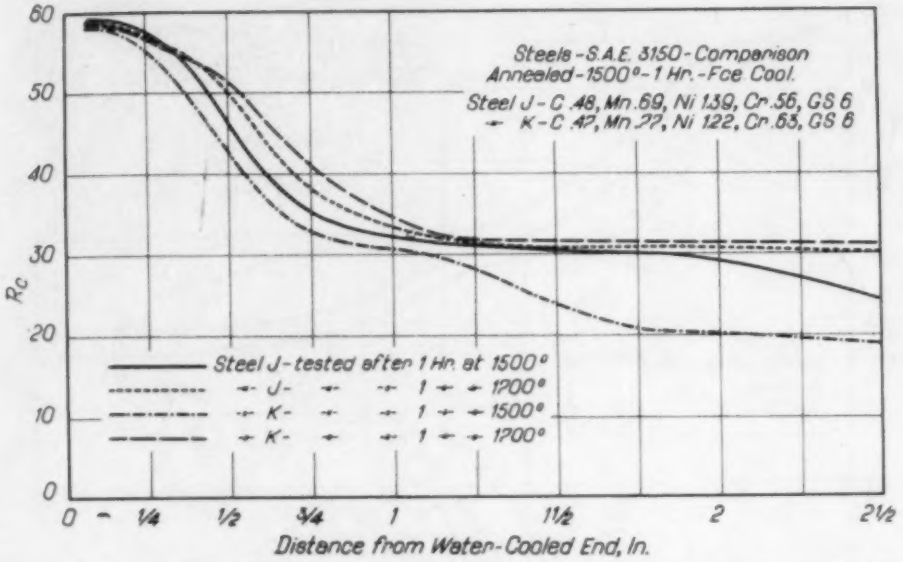
Steels - S.A.E. 3150 - Comparison
Annealed - 1500° - 1 Hr. - Fce Cool.
Steel J - C .48, Mn .69, Ni 1.39, Cr .56, GS 6
" K - C .47, Mn .77, Ni 1.22, Cr .63, GS 6
Steel J - tested after 1 Hr. at 1500°
" J - " " 1 " " 1200°
" K - " " 1 " " 1500°
" K - " " 1 " " 1700°
Rc
Distance from Water-Cooled End, In.

**Fig. 14—Variation in Time-Temperature Requirements for Attaining "Maximum" End-Quenched Hardenability.**

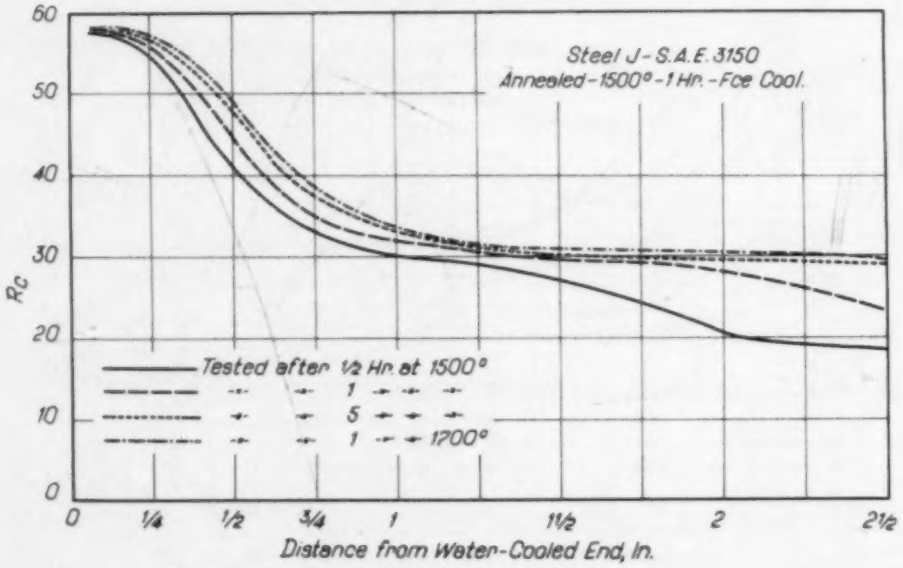


**Fig. 15—Variation in Time-Temperature Requirements for Attaining "Maximum" End-Quenched Hardenability.**

Fahr. was cooled in the furnace to 1200 degrees Fahr. before testing.

The behavior of five typical heats of S.A.E. 4150 is summarized in Fig. 19. The response of these steels clearly varies as did S.A.E. 3140. Here again is an indication of the rate of carbide solution; although the Cr-Mo carbide has always been considered sluggish, "maximum" hardenability was achieved after ½ hour at 1550 degrees Fahr. on two of these steels (the greater effectiveness of 1550 degrees Fahr. as compared to 1500 degrees Fahr. is also a factor).

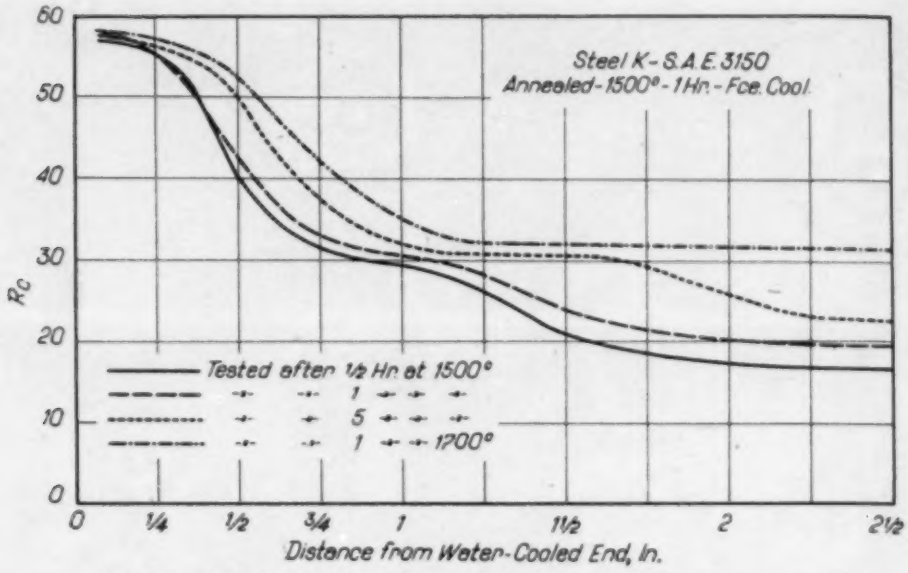


Fig. 16—Variation in Time-Temperature Requirements for Attaining "Maximum" End-Quenched Hardenability.

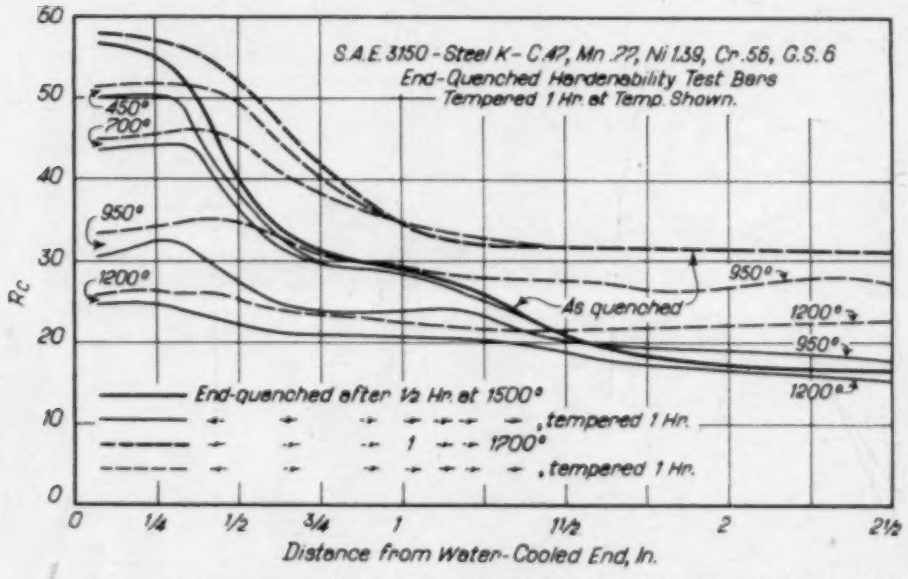


Fig. 17—Tempering Versus Completeness of Hardening S.A.E. 3150 Steel.

It may well be that this carbide is slower in dissolving, but that homogeneity is normally better for this type of steel. Microstructure of annealed samples failed to reveal any such differences in banding as were found in the S.A.E. 3140 samples shown in Fig. 14. It is notable that three of these S.A.E. 4150 heats show better hardenability than the lowest S.A.E. 4340 of Fig. 18.

In consequence of the above and many other series of tests, a standard procedure has been adopted by the author as follows: Two

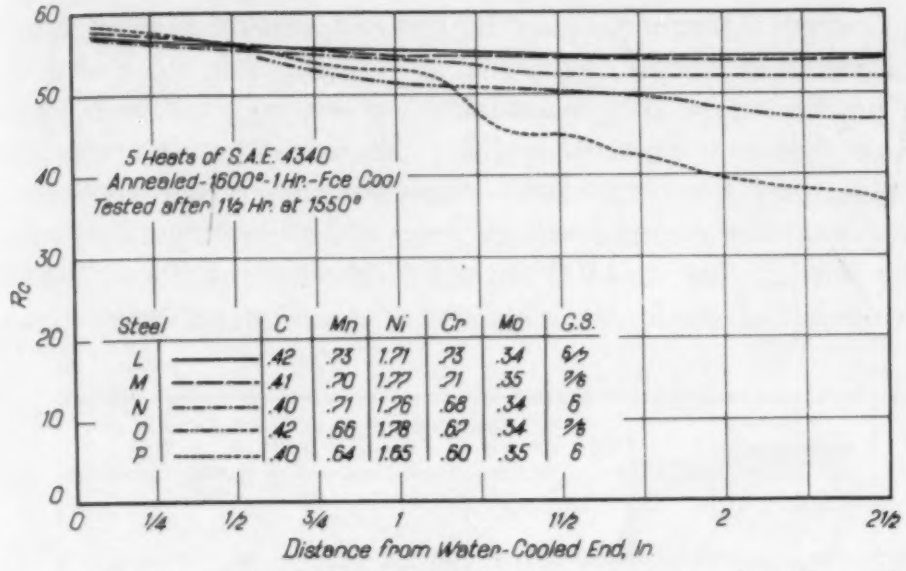
5 Heats of S.A.E. 4340
Annealed-1600°-1 Hr.-Fce Cool
Tested after 1½ Hr. at 1550°
Rc
Steel | C | Mn | Ni | Cr | Mo | G.S.
L | .42 | .73 | 1.71 | .73 | .34 | 6-7
M | .41 | .70 | 1.77 | .71 | .35 | 7/8
N | .40 | .71 | 1.76 | .68 | .34 | 6
O | .42 | .66 | 1.78 | .62 | .34 | 7/8
P | .40 | .64 | 1.65 | .60 | .35 | 6
Distance from Water-Cooled End, In.

Fig. 18—End-Quenched Hardenabiiity of Five Heats of S.A.E. 4340 Steel.

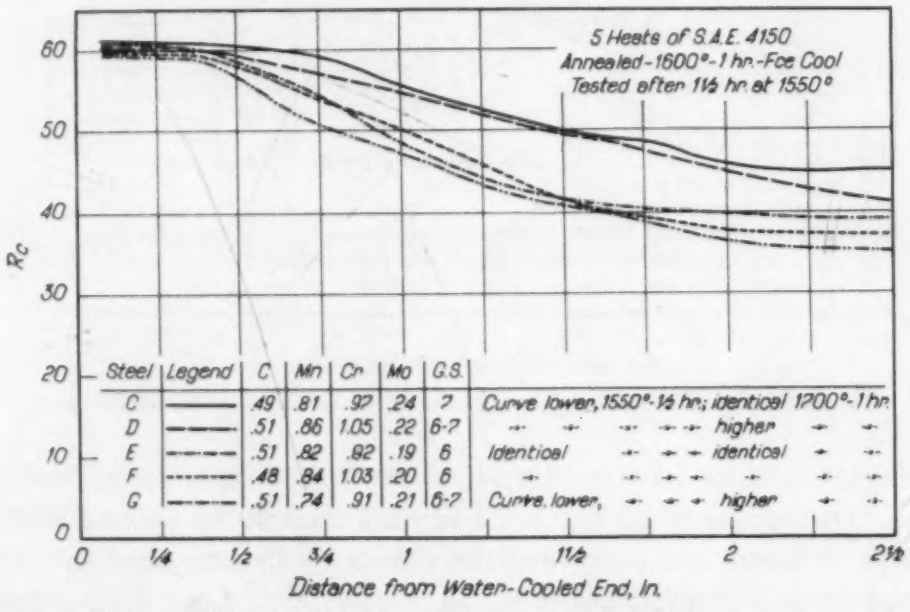
5 Heats of S.A.E. 4150
Annealed-1600°-1 hr.-Fce Cool
Tested after 1½ hr. at 1550°
Rc
Steel | Legend | C | Mn | Cr | Mo | G.S.
C | .49 | .81 | .92 | .24 | 7 | Curve lower, 1550°-½ hr; identical 1200°-1 hr.
D | .51 | .86 | 1.05 | .22 | 6-7 | higher
E | .51 | .82 | .92 | .19 | 6 | Identical identical
F | .48 | .84 | 1.03 | .20 | 6 | higher
G | .51 | .74 | .91 | .21 | 6-7 | Curve lower,
Distance from Water-Cooled End, In.

Fig. 19—End-Quenched Hardenability of Five Heats of S.A.E. 4150 Steel.

specimens are prepared from the annealed steel to be checked, the first being end-quenched after a normal time at heat for that steel, and the second tested at either shorter or longer time, as suggested by the type of hardenability curve obtained from the first. If the normal test gives a curve flat throughout the lower hardness range, shorter time is used for the second test. If, however, the first curve indicates less than "maximum" hardenability by a secondary falling

of the curve, a longer time or higher temperature is used for the second test.

A peculiar type of hardenability curve was manifested by two steels, as shown in Figs. 20 and 21. While neither is a commonly used type, they are occasionally employed for certain applications. The hardenability curves resulting from end-quenching after a short or long heating time at 1500 or 1600 degrees Fahr. show "humps" in hardness near Rc 35-40; a definite increase in hardness amount-

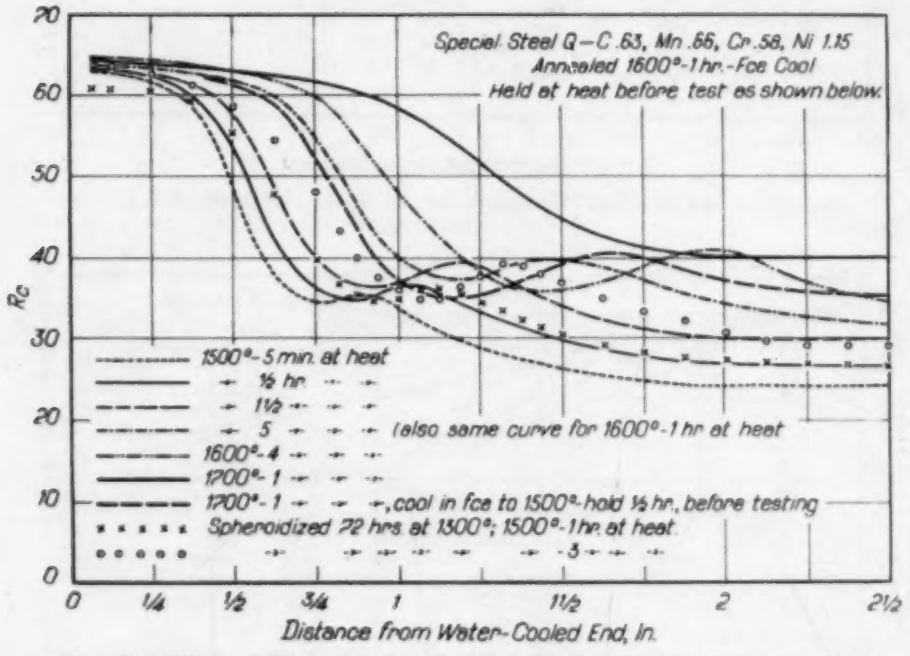


Fig. 20—Effect of Time and Temperature on End-Quenched Hardenability.

ing to as much as 5 points Rc is found at a distance of 1 inch or more from the quenched end, well beyond the end of the martensitic zone. This hump moves toward the slower-cooled end, and increases both in level and in magnitude, as the holding time is increased. However, when the steels were treated after 1 hour at 1700 degrees Fahr., the hump was not produced (within the length of the specimen). On Steel Q, however (Fig. 20), dropping from 1700 to 1500 degrees Fahr. before testing again gave the hump.

To check on the effect of carbides definitely out of solution, two specimens of each of these steels were spheroidized for 72 hours at 1300 degrees Fahr., then heated at 1500 degrees Fahr. for 1 hour and 3 hours before testing. The humps were still revealed, although micro-examination showed clearly that carbides had not been com-

pletely redissolved in either test. This slow rate of solution points to considerable sluggishness, undoubtedly a factor in failure to homogenize. The specimen quenched from 1700 degrees Fahr., with consequent avoidance of the hardness hump, was still decidedly banded; therefore it may be considered that whatever peculiar behavior occurred from the lower temperatures was merely suppressed or moved off the curve, not eliminated.

The metallographic structures found associated with these hard-

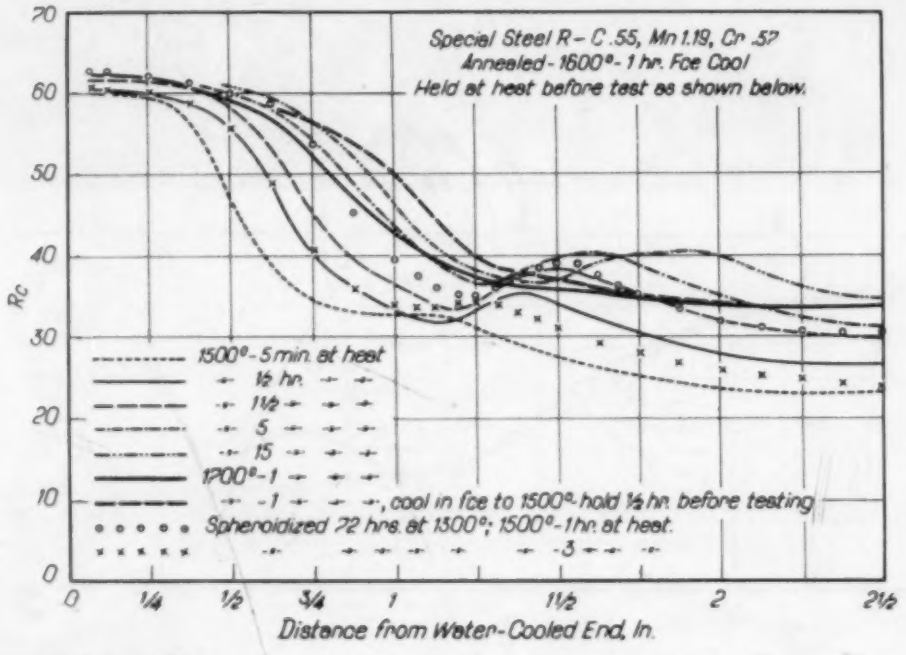
Special Steel R – C .55, Mn 1.19, Cr .57
Annealed - 1600° - 1 hr. Fce Cool
Held at heat before test as shown below.
Rc
1500° - 5 min. at heat
½ hr.
1½
5
15
1700° - 1
-1
, cool in fce to 1500° - hold ½ hr. before testing
Spheroidized 72 hrs. at 1300°; 1500° - 1 hr. at heat.
-3
Distance from Water-Cooled End, In.

Fig. 21—Effect of Time and Temperature on End-Quenched Hardenability.

ness increases near Rc-35 were interesting. In steel R, which was severely banded, the transition zone of hardness from Rc-60 to Rc-35 consisted of martensite with increasingly wide bands of "pseudo-martensite" (acicular troostite or bainite). Martensite persisted in bands even beyond the peak of the hardness humps, but at the secondary hardness peaks the bands which had been "pseudo-martensite" gradually became pearlite. Some of these structures are shown in Fig. 22.

It has been shown, particularly by Griffiths, Pfeil and Allen (8), that "many alloy steels differ from carbon steels in that, when cooled continuously at such a rate that the pearlite point is suppressed, they do not always give martensite or mixed martensite and austenite, as carbon steels do, but may transform in the intermediate

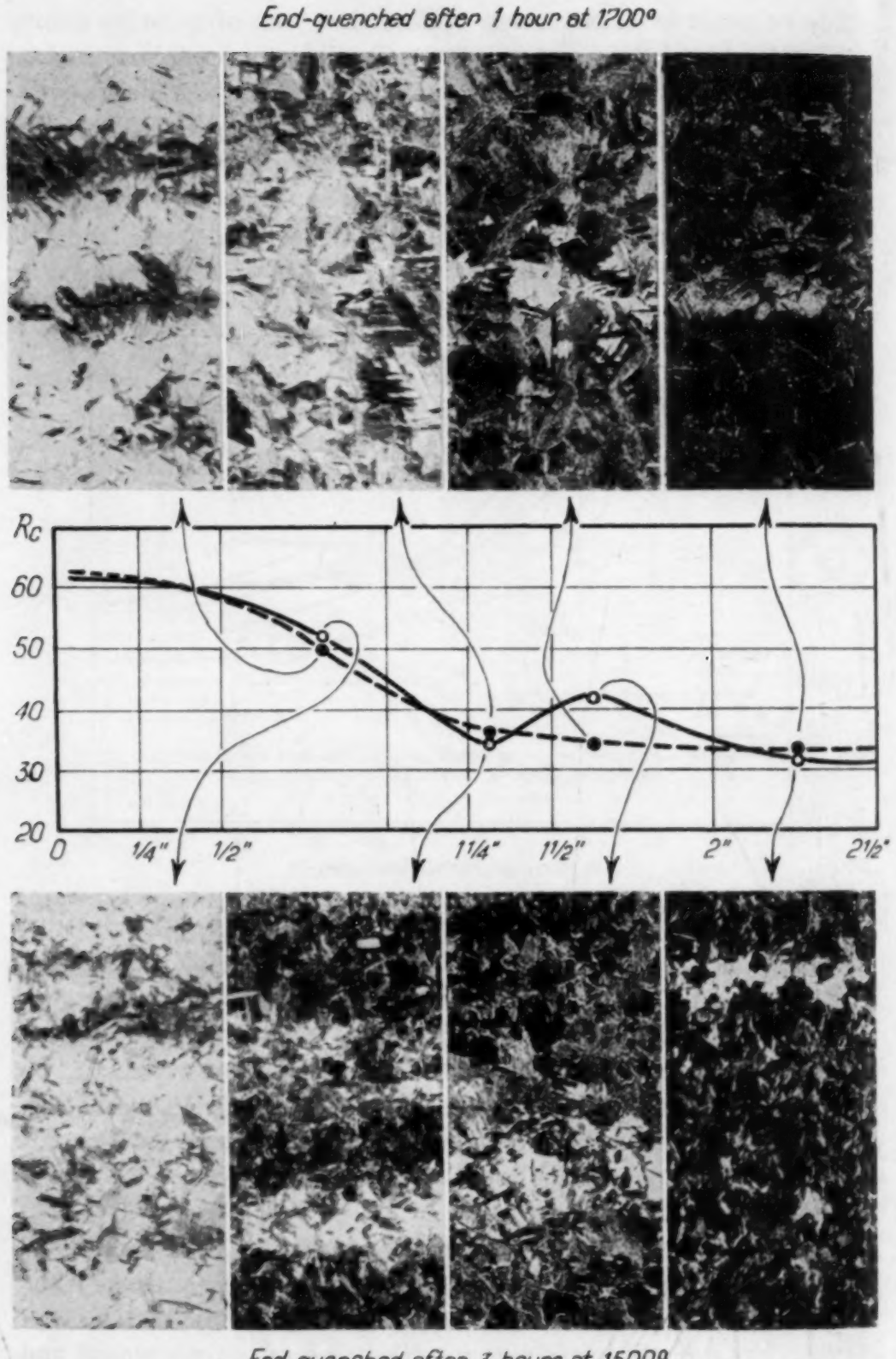


Fig. 22—Microstructure of Steel R With and Without Hump. × 150.

temperature range between 1020 and 480 degrees Fahr. (550 and 250 degrees Cent.)" (8). Clearly, steel R is of this type. Transformation of samples of steel R by quenching in lead at 750, 850, 1000, and 1100 degrees Fahr. showed on micro-examination that after 30

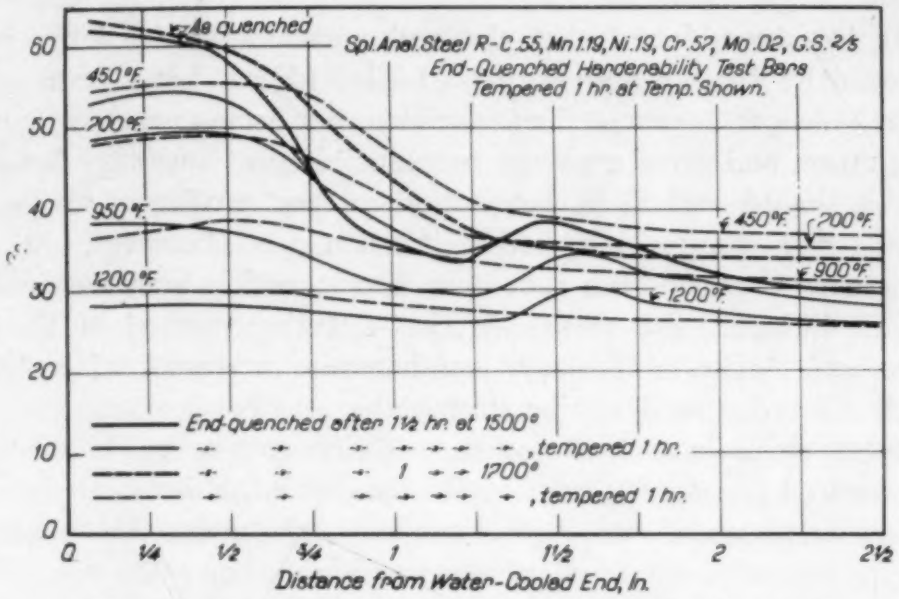


**Fig. 23—Tempering Versus "Hump" in Hardenability Curve.**

seconds, approximately equal amounts of transformation constituents (estimated at 10 per cent) had formed at 850 and 1100 degrees Fahr., while 60 seconds were required to develop similar amounts at 750 or 1000 degrees Fahr., indicating a different type of curve than the customary "S" curve for eutectoid carbon steels.

Perhaps this mode of transformation, coupled, no doubt, with the peculiarities resulting from banding, can be the cause of the secondary hardness increase shown on the hardenability curves in Figs. 20 and 21. In Fig. 23 are shown the results of tempering two of the specimens, one of which revealed the hardness hump, the other in which this had been suppressed by end-quenching from 1700 degrees Fahr. Far greater retention of austenite seems to be indicated by action of the specimen from the higher temperature treatment.

This "hump" in the hardenability curve at lower hardnesses has occasionally been found to the extent of one or two points Rockwell, and has been noted in other laboratories (9), but these two steels are the only ones the writer has encountered which show such a marked increase. It appears probable that the "hump" is simply

an extreme condition of what is found in incompletely homogenized samples, which usually show a temporary flat in the curve followed by a further decrease.

## Conclusion

It has been found that the end-quenched hardenability test developed by Jominy and Boegehold (2), (3), is not critical with respect to length or diameter of specimen, nor as to quenching water temperature, and gives excellent reproducibility of results. Accordingly, it should find rapid acceptance by steel producers and consumers for checking of alloy steels. Carbon steels, however, produce such a narrow martensitic zone that discrimination between similar steels is difficult. For these, the S-A-C rating method of Burns, Moore and Archer (1), based on hardness traverses of a fully quenched 1-inch round, can be employed.

Alloy steels have been found to differ greatly in the time at heat required for development of the maximum hardenability which the steel can attain below the grain growth temperature, this variation not being related to the standard chemical analysis or grain size. For each steel there is a different minimum requirement in time-temperature conditions of holding at heat before hardening, if this "maximum" hardenability is to be achieved. The differences in hardenability among steels of the same type may be greater than those from type to type after ordinary heating times. Thus, the purpose of selection of one steel over another may be defeated, and extra price wasted. Jominy (3) showed curves representing S.A.E. 3145 and 5145 samples, the latter falling off from 1¾-inch on, so that the two curves crossed after being parallel for some distance. It would appear that this is another example of the sort discussed above, and had the S.A.E. 5145 been held sufficiently long at heat, it would have given a continued flat plot throughout the lower hardness range.

It is suggested that homogeneous austenite before quenching is required for maximum hardenability, and that degree of banding is a possible indication of the ease or difficulty with which this will be attained. A recently developed deoxidizer (Grainal) has been found to give fine grain size, coupled with intense hardenability (10). May it not be that the effect of this Al-V-Ti treatment is to promote homogeneity? The strong normality of steels so treated would seem to indicate this. For homogeneity of austenite, it is necessary that carbides diffuse into and through zones of low carbon content (such

zones being obviously more resistant to carbon diffusion). Other elements as well must diffuse. It will undoubtedly be found possible to make steel which will attain "maximum" hardenability quickly and economically, with only small percentages of alloying elements, the steel being so constituted as to insure best usage of the alloys, by regulation of factors opposing diffusion and homogenization.

It is hoped that the results reported herein may lead to increased interest in and use of the end-quenched hardenability test, and that data thus found will be disclosed, so that in time, a better understanding of the factors involved in hardenability may be had. The ultimate practical object is to be able to make, buy, and use steels of known and uniform hardenability.

The author does not think or wish to imply that hardenability is the only important feature of alloy steel behavior. However, uniform hardening response is favorable to production of steel parts of uniform properties, and this, and most economical use of steel, will result when hardenability is uniform or more nearly so than today. Maximum hardenability may not be desired. There are many cases in which shallow hardening is deliberately produced by heating cycles that leave carbides out of solution, but uniformity is still most desirable.

Effect of rate of heating, of actual quenching temperature, and of variations within single commercial heats of alloy steels must be studied in reference both to hardenability and rate of attainment of "maximum" hardenability.

In conclusion, the author wishes to acknowledge the assistance of W. A. Silliman, who performed many of the tests, as well as metallographic work; also, the continued interest of H. W. McQuaid, whose stimulating comments and advice led to much of the work reported here. Thanks are also due to The Republic Steel Corporation for several of the chemical determinations listed in Table I. The balance of the work was carried on in the metallurgical laboratory of the Cleveland Tractor Company.

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## DISCUSSION

**Written Discussion:** By W. E. Jominy, Research Laboratories Division, General Motors Corp., Detroit, Mich.

Naturally, I am very pleased with this paper using the end-quench method for studying hardenability and especially with the conclusion that it gives excellent reproducibility of results. This confirmation of our own work is, of course, gratifying.

The author's observation that the method is not so well suited to plain carbon steels is in accordance with our own observations and it was for this reason that a modified test bar was developed for the shallow hardening steels. This modified test bar, which is described at this convention, gives good differentiation between shallow hardening steels as well as very satisfactory reproducibility of results.

In determining the precise effect of variation of bar length and water temperature the author has obtained needed data. The fact that these conditions are not critical aids the easy working of the test. However, it is to be hoped that variations as wide as from 35 to 140 degrees Fahr. are not permitted in making the test since this would cause variations in results obtained with certain steels. In my experience some steels will show a measurable difference, though small, when the cooling water temperature is changed from 70 to 80 degrees Fahr. I would consider it desirable to keep the cooling water between 73 and 78 degrees Fahr. This is very easily done.

One more word of caution in making the test concerns the depth below the surface of the bar at which hardness readings are made. As stated by the author, these readings can be made at the surface, if there is no decarburization, or at any depth below the surface. There is an exception to this, however, and that is with carburized test bars. In this case readings should be taken at 0.015 inch below the surface.

The author's finding of the variation between five heats of S.A.E. 3140 steel all given identical treatment brings up the old argument about predicting hardenability from chemical composition. Certainly the wide differences in hardenability are not accounted for by the small differences in the compositions as given by the author. The effect of increasing heating time is quite

interesting. Of course, an increase in hardenability due to increased grain size resulting from heating a longer time or at a higher temperature is expected but the author's results seem to show a change in hardenability far out of proportion to this expected change. I have experienced this same effect of time of heating with spheroidized high carbon-plain carbon steels and the results are given in my paper at this convention.

The author's belief that this effect is the result of homogenizing the structure is shown, he says, by the decrease in banding as "maximum" hardenability is approached. Could the author tell us whether this homogenizing treatment has a permanent effect on the microstructure or if when the steel is reannealed will the banded structure again appear?

The author is to be congratulated for this thought provoking paper.

**Written Discussion:** By H. W. McQuaid, assistant chief metallurgist, Republic Steel Corp., Cleveland.

This paper is of the greatest interest to me and, I believe, contains food for a great deal of thought for both the producer and user of steel. In the first place it indicates the value of the Jominy end-quench test in evaluating alloy steels, especially those of the medium carbon grade.

Figs. 9a and 9b are worthy of considerable thought as they indicate that the character of the austenite at the time of quenching is an important factor in hardenability results. A heterogeneous condition of the austenite, as regards carbon diffusion at the time of quenching, would result in lowered hardenability. If the steel is of a type in which the crystal boundaries resist carbon diffusion due to the presence of other elements in solution there is a possibility that the grain boundaries are not only low in carbon but in some cases it is believed are practically carbon-free. Thus, in spite of the best of quenching conditions, we find that the grain boundaries consist of low hardness transformation constituents and in some cases are completely ferrite.

Mr. Williams has shown that there is considerable variation in the time and temperature at which full hardenability is obtained. It is interesting to note that in steels of commercial equivalent grain size one steel will reach full hardenability in a few minutes at 1500 degrees Fahr. whereas in a steel of corresponding analysis and grain size 8 hours at 1700 degrees Fahr. was insufficient to increase the hardenability to its maximum for this steel. The curves indicate that for sections up to approximately 1 inch, the steels shown in this paper do not vary beyond commercial limitations except possibly steel "X", but that in heavier sections the difference in hardenability between the heats shown is very marked unless sufficient temperature and time is given to insure a more uniformly stable austenite.

**Written Discussion:** By M. C. Fetzer, instructor in metallurgy, The Pennsylvania State College, State College, Pa.

Mr. Williams has stated that microstructural heterogeneity as evidenced by banding is decreased as "maximum hardenability" is produced with increased time or temperature. This is a valuable contribution. In the main, however, it seems that the author has not given sufficient weight to the effect of grain size, especially fracture grain size. It is assumed that if the McQuaid-Ehn test shows a fine grain that all heat treatments less severe than 8 hours at

1700 degrees Fahr. will produce the same fine grain size and that the attainment of "maximum hardenability" is due to homogenization.

In my own experience, I find that the fracture grain size increases with time and temperature even below the "coarsening" temperature (if the steel does have a "coarsening" temperature). Although steels of "similar" composition may have different hardenabilities even for the same grain size, increasing the grain size for a given steel usually increases the hardenability. Mr. Williams varied his times and temperatures from 5 minutes to 3 hours at 1500 degrees Fahr. to 1 hour at 1700 degrees Fahr. and attributed the change in hardenability to homogenization. It would have been enlightening to have had the fracture grain sizes after these varying soaking treatments.

It is to be expected that the hardness curve will tend to level out to a horizontal line as the grain size is increased because when a certain grain size is attained the reaction rate will be slow enough so that all of the pearlite formed will have been formed at the knee of the "S" curve. (An increase in grain size shifts the curve to the right.)

### Author's Reply

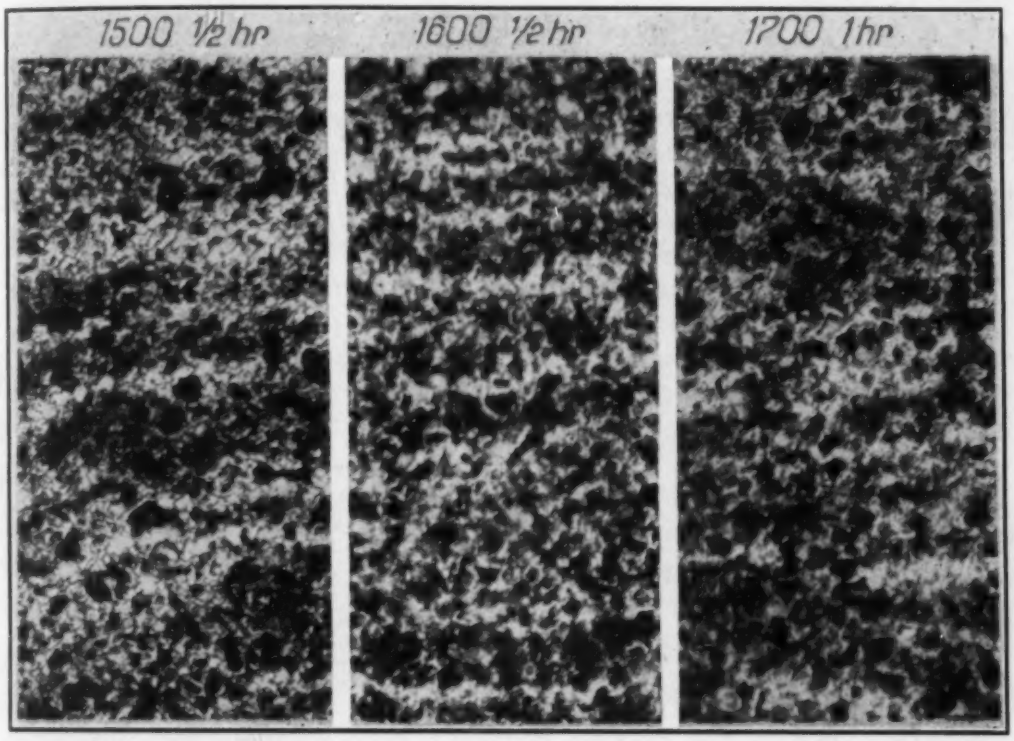
The author appreciates the comments offered. The discussion by Mr. Jominy is especially welcome, since he as originator of the test method is to that extent responsible for the appearance of this paper. He points out that he has found certain steels to be sensitive to water temperature. In no sense did the author wish to recommend that quench-water temperature be permitted to vary between 35 and 140 degrees; this was normally kept within the range of 70-75 degrees as Jominy indicates is preferable.

In response to his question as to permanence of decrease in banding as "maximum" hardenability is approached, the microstructures herewith are presented. Samples of one steel after three different hardening treatments which gave different end-quench hardenability curves (see text) were annealed at 1700 degrees Fahr. for one-half hour, then sectioned and examined. As shown here at 100 diameters, there appears to be a decrease in banding as hardening temperature was raised. However, it is possible that a longer annealing time might have brought all back to the original conditions. Probably, whatever promotes banding is suppressed rather than removed by longer time or higher temperature.

It might well be mentioned that Mr. Jominy has checked the S.A.E. 5140 referred to in the conclusion of the present paper, and finds that its hardenability curve did improve with longer heating time, as suggested.

The remarks by Mr. McQuaid on austenite homogeneity are pertinent and worthy of note. As he also points out, the major differences in hardenabilities of the individual steels of the various types are principally those to be expected in sections above 1 inch or more; little differences would thus be found in small sizes; however, as remarked in the paper, larger sizes may reflect these variations to an important degree. Further, we may consider these differences to be the result of some variable which must ultimately be controlled if we are to have uniform steels of predictable behavior; this factor may be even more significant as it affects the ease with which steels may be annealed by direct transformation, a matter of increasing interest.

We must bear in mind the fact pointed out by Dr. Fetzer that actual grain size changes even below the coarsening temperature. It is quite likely that this has an effect; unfortunately, fracture or other grain size tests were not run concurrently. The correlation with grain size is thus not shown; however, from additional tests run at temperatures up to 2000 degrees Fahr., the author has come to believe that grain size is most important as a factor in formation of homogeneous austenite. This, however, is certainly not proved. The high hardenability conferred by Grainal treatment, which involves only a slight alloy addition, may be considered to indicate that degree of normality is much more significant than grain size, since this treatment produces a fine-grained steel. It is to be hoped that Dr. Fetzer will soon publish some of the results obtained with the end-quench test in his experimental work, which it is understood has been extensive.



# AN ANALYSIS OF MACHINABILITY DATA FROM COLD FINISHED AND HEAT TREATED S.A.E. 1045 STEEL

By O. W. Boston and W. W. Gilbert

*Abstract*

*This paper presents the results of a series of tests showing the relation of physical properties and microstructure to machinability for several logs of a medium carbon steel, S.A.E. 1045. The machinability data include the cutting speed–tool life relation, cutting temperatures, cutting forces, and surface quality when dry cutting a test log of each of five structures; the tool life–cutting speed relation when turning the cold finished log with different types of cutting fluids; and the surface quality when turning the cold finished log at different speeds.*

## The Material Cut

SIX logs of 3-inch diameter cold-finished S.A.E. 1045 steel were prepared as indicated in Table I. The carbon content of this steel was 0.41 per cent. The balance of the analysis was manganese 0.64 per cent, silicon 0.23 per cent, sulphur 0.028 per cent and phosphorus 0.027 per cent.

After running the various tests, a standard tensile test specimen was turned from each log and the physical properties were determined, as recorded in Table I. Photomicrographs giving the structure of each specimen at the 2.5-inch diameter are shown in Figs. 1, 2, and 3 at magnifications of 100, 500, and 1000 diameters respectively. These microstructures show rather definite changes between the various bars, as follows:

Log 1 shows a microstructure of rather coarse-grained pearlite with heavy ferrite boundaries.

Log 2 shows a fine-grained nodular troostitic-sorbitic structure and represents one of the two hardest specimens tested. It has a Rockwell "B" hardness of 102.3 near the outer surface.

Log 3 shows no distinct change in the metallographic structure from that of Log 2. The Rockwell "B" hardness remains at 103.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, O. W. Boston is professor of metal processing, and W. W. Gilbert is assistant professor of metal processing, University of Michigan, Ann Arbor, Michigan. Manuscript received June 19, 1939.

**Table I**

**Physical Properties of the Six S.A.E. 1045 Steel Logs**

| Log Number | Treatment of the Log | Yield Point (A) p. s. i. | Ultimate Strength p. s. i. | Per Cent Elongation | Per Cent Reduction of Area | Rockwell "B" Hardness (b) | Rockwell "B" Hardness (c) | $V_1$ (d) f.p.m. | $V_1$ (d) Per Cent |
|---|---|---|---|---|---|---|---|---|---|
| 1 | Cold-finished | 89,800 | 92,850 | 18.75 | 42.70 | 91.3 | 96.8 | 192 | 168 |
| 2 | Oil-quenched from 1520 deg. F. | 64,200 | 104,250 | 22.65 | 50.00 | 89.5 | 102.3 | 114 | 100 |
| 3 | Oil-quenched at 1520 deg. F. and drawn at 600 deg. F. | 62,150 | 103,100 | 22.65 | 48.55 | 92.5 | 103.0 | 127 | 111 |
| 4 | Oil-quenched at 1520 deg. F. and drawn at 800 deg. F. | 63,650 | 104,200 | 22.65 | 50.00 | 92.0 | 101.0 | ... | ... |
| 5 | Oil-quenched at 1520 deg. F. and drawn at 1000 deg. F. | 57,600 | 98,050 | 25.00 | 51.90 | 91.5 | 97.9 | 141 | 124 |
| 6 | Annealed at 1600 deg. F. | 59,300 | 79,100 | 27.30 | 41.20 | 78.7 | 82.7 | 205 | 180 |

Log 1 was cold-finished to three-inch-diameter as received.
Log 2 was the three-inch-diameter cold-finished bar, oil-quenched from 1520 degrees Fahr.
Logs 3, 4, and 5 were quenched in oil from 1520 degrees Fahr. and drawn for two hours as indicated.
Log 6 was the three-inch-diameter cold-finished bar, annealed by furnace cooling after holding at 1600 degrees Fahr. for three hours.
(a) Standard tensile test specimens were machined from the center of the three-inch logs used in the force and tool-life tests.
(b) Average of three tests on a flat on large end of tensile test bar.
(c) Average of tests of 2.5-inch diameter of the heat treated disks ½-inch thick in the facing tests sawed from the end of the logs.
(d) $V_1$ is the cutting speed in f. p. m. for a tool life of one minute. The per cent rating is based on the quenched bar No. 2 as 100 per cent for $V_1$ of 114 f. p. m.

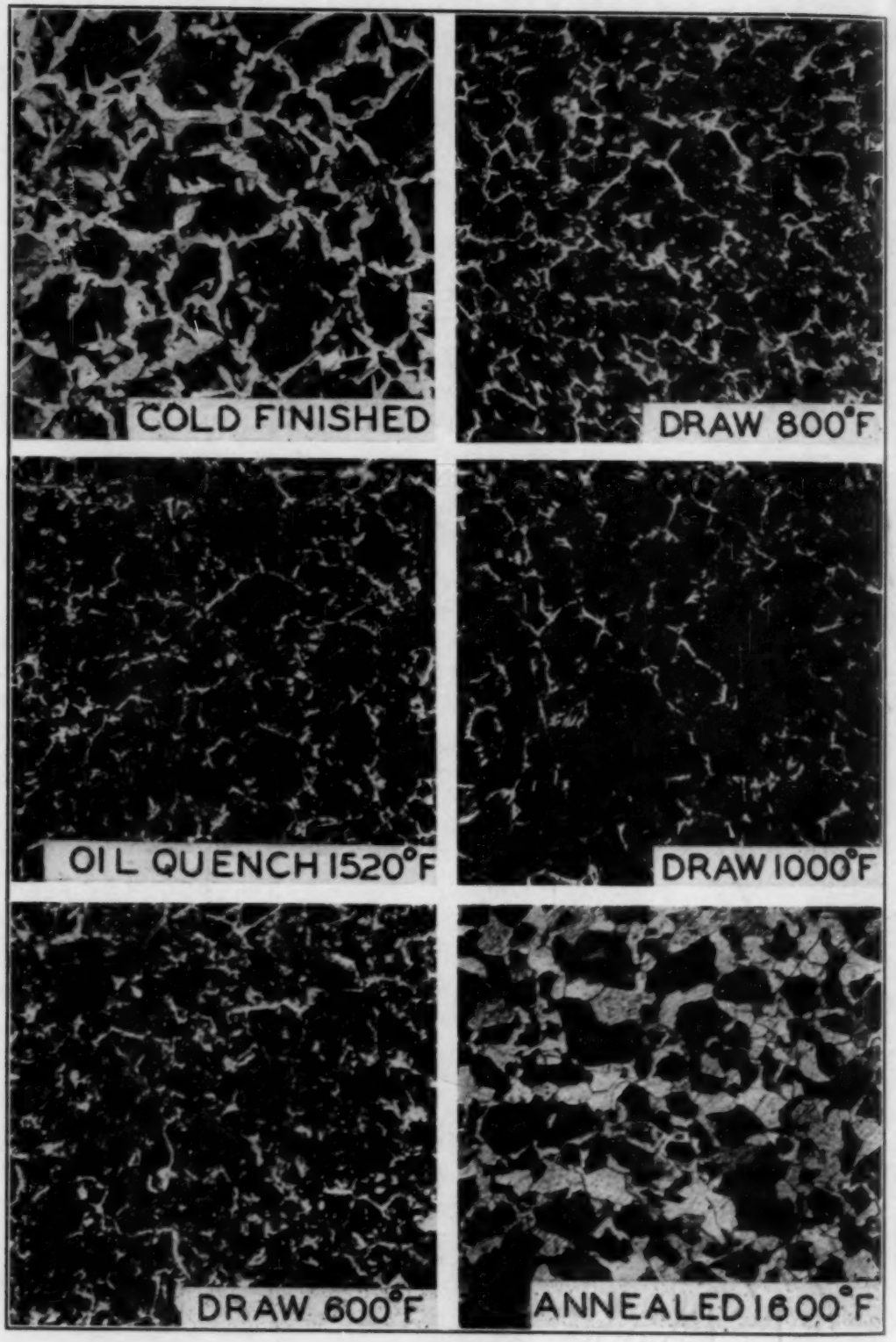



Fig. 1—Microstructure of S.A.E. 1045 Steel. × 100.

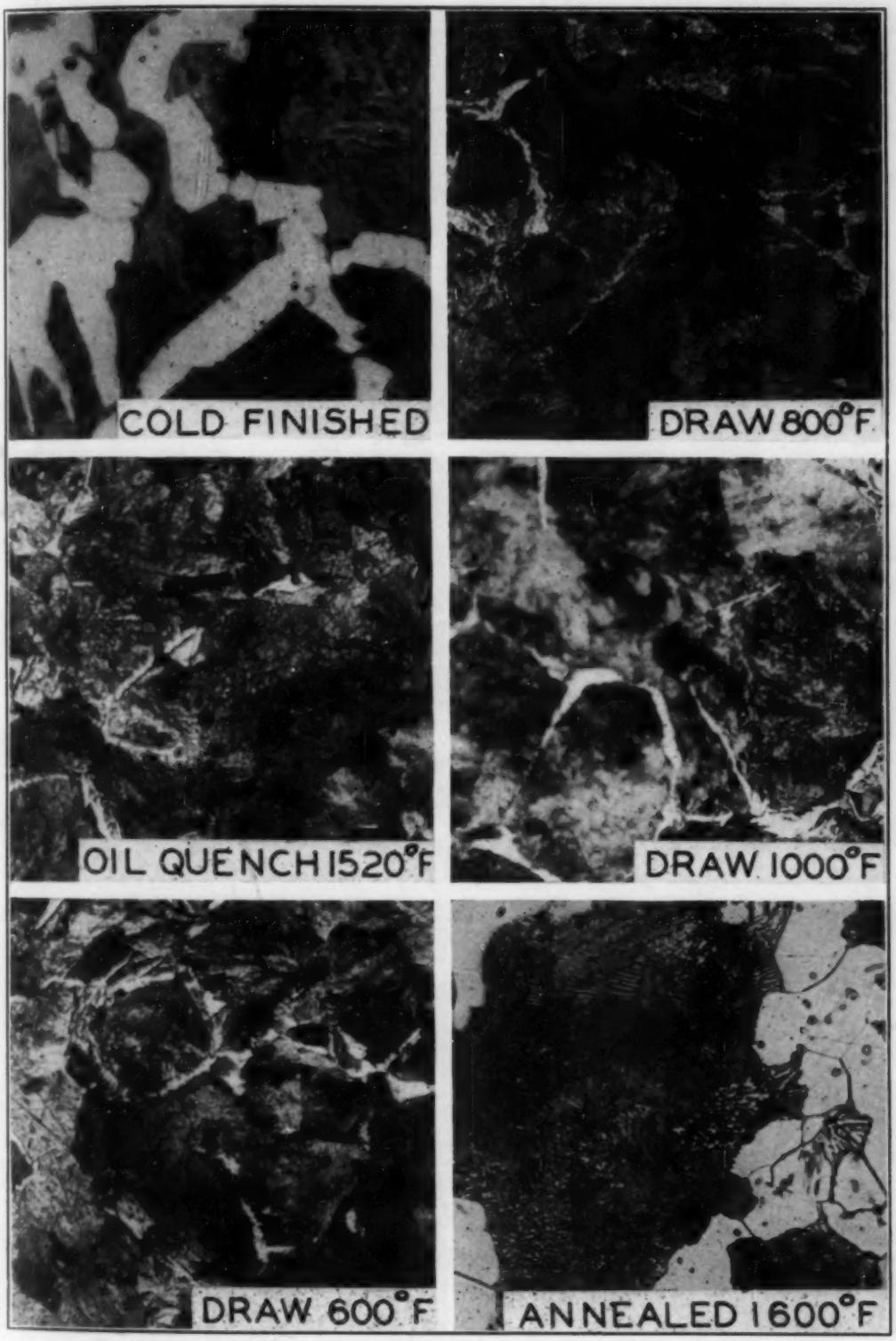



Fig. 2—Microstructure of S.A.E. 1045 Steel. × 500.

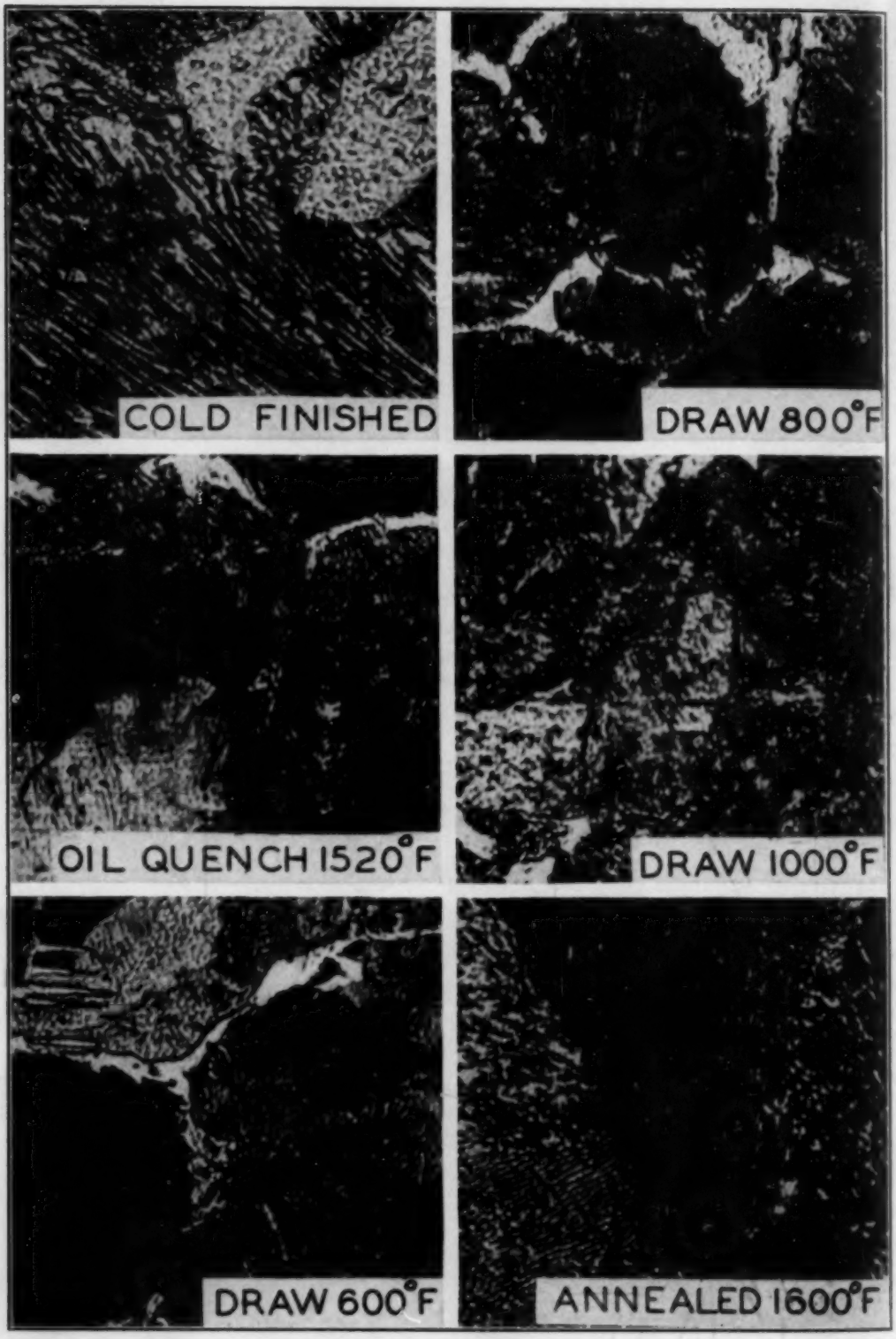



Fig. 3—Microstructure of S.A.E. 1045 Steel. × 1000.

Log 4 shows a troostitic-sorbitic structure similar to that of Log 3. Less troostite is present, and more sorbite is evident, and the grain boundaries (ferrite) are slightly heavier.

Log 5 indicates that a further transformation has taken place because of the higher drawing temperature. The troostite has been reduced considerably.

Log 6 shows a characteristic annealed structure of ferrite and lamellar pearlite.

## Cutting Speed–Tool Life Tests

Using a heavy-duty engine lathe provided with an infinitely variable speed drive, tool life values were determined when turning each bar at several different speeds. The cutting tool, of the 18-4-1 type of high speed steel in the form of ⅜-inch tool bits, was ground to a shape of 8-degree back rake, 22-degree side rake, 6-degree side and end relief, 6-degree end-cutting edge angle, 15-degree side-cutting edge angle, and $\frac{3}{64}$-inch nose radius. Its symbol is 8-22-6-6-6-15-$\frac{3}{64}$. The tool bits were supported in a rigid holder.[1] All tests were run dry, and the depth of cut was kept constant at 0.10-inch and the feed at 0.0127-inch per revolution of the work.

The results of these tests are shown plotted on log-log paper in Fig. 4. Straight lines have been drawn through the experimental points; as in many other tests, this condition almost invariably maintains. Each of these lines is represented by an equation corresponding to the form $VT^n = C$, in which V is the cutting speed in feet per minute, T is the tool life in minutes up to complete tool breakdown, $n$ is the slope from the horizontal of the straight line on log-log paper, and C is the constant depending upon the tool, cutting fluid, size of cut, and material. In this case, the material was the only variable. Values of $n$ are found to be 0.11 in all cases, indicating that all five lines are parallel. The values of C, which represent the cutting speed for a tool life of one minute, are shown on the graph to be as follows: With the value of C for the bar No. 2 taken as 100 per cent, the cutting speed ratings for the other bars would be as indicated in per cent as follows:

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[1] O. W. Boston, W. W. Gilbert, and C. E. Kraus, "Factors Influencing the Nature of the Cutting Speed-Tool Life Curve," Transactions, American Society for Metals, Vol. 24, March, 1936, p. 186. (Tool *A*. Table I.)

| | $V_1$ in f.p.m. | Rating |
|---|---|---|
| Bar 2—(oil-quenched) | 114 | 100% |
| Bar 3—(oil-quenched and drawn at 600 degrees Fahr.) | 127 | 111 |
| Bar 5—(oil-quenched and drawn at 1000 degrees Fahr.) | 141 | 124 |
| Bar 1—(cold-finished) | 192 | 168 |
| Bar 6—(annealed) | 205 | 180 |

Corresponding values of tool life in minutes for a cutting speed of 130 feet per minute would be

| | | |
|---|---|---|
| Bar 2—0.3 | Bar 5—2 | Bar 6—64 minutes |
| Bar 3—0.8 | Bar 1—37 | |

This shows that the annealed structure of ferrite and pearlite grains of about the maximum refinement for this steel gives the highest cutting speed for a given tool life. The coarse-grained pearlite with heavy ferrite boundaries of the cold drawn specimen

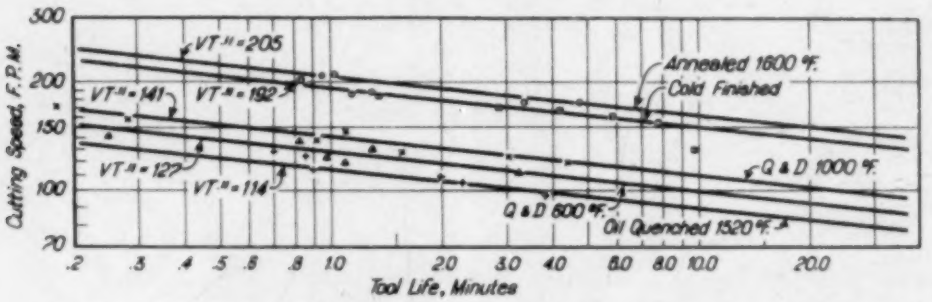


Fig. 4—Tool Life-Cutting Speed Relations When Turning Annealed, Cold-Finished, and Heat Treated S.A.E. 1045 Steel, Dry. Depth of Cut, 0.100 Inch, and Feed, 0.0127 Inch. ¾-Inch-Square High Speed Steel Tools of 8-22-6-6-6-15-¾ Shape Were Used.

gives a cutting speed approaching that of the annealed specimen. The lowest cutting speed is obtained for Bar 2, the oil-quenched specimen, having the highest hardness and showing a fine-grained nodular troostitic-sorbitic structure. While drawing at 600 degrees Fahr. has not produced a very distinct change metallographically, an improvement in cutting speed of 11 per cent is obtained. By drawing at 1000 degrees Fahr., the troostite is transformed to sorbite, and a definite improvement in cutting speed of 24 per cent is obtained. The cutting speed-tool life line for the steel oil-quenched and drawn at 800 degrees Fahr. is not shown, but it falls between that for the 600 and 1000 degrees Fahr. draw lines, and indicates that, as the troostite is gradually transformed to sorbite, a corresponding improvement in cutting speed is obtained.

Cutting speed-tool life tests were run on the cold-finished log

with several types of cutting fluids, such as dry cutting, an emulsion, a sulphurized-mineral oil, and a sulphurized-base oil. By the time one cutting speed-tool life line had been obtained, the bar was reduced in diameter to such an extent that the physical properties were changed and no satisfactory results could be obtained. In other words, there was insufficient uniform material to give uniform test points for the four cutting fluids tested.

## Cutting Forces

Tangential forces were determined when turning each of the heat treated bars when using the simple tool dynamometer.[2] A ¼-inch square high speed steel tool, held in the dynamometer, had a shape of 8-22-6-6-6-15-.01, the same as that used in the tool-life tests. Forces for each bar were obtained, first, by keeping the depth

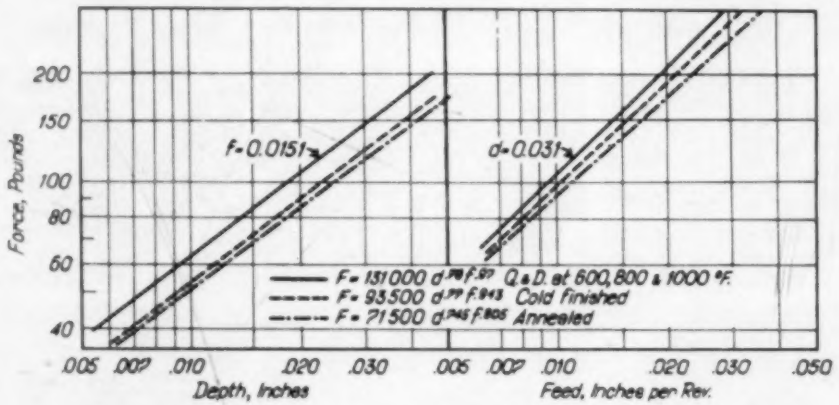


Fig. 5—Tangential Forces on Cold-Finished, Annealed, and Heat Treated Bars of S.A.E. 1045 Steel, Using a Tool Shape of 8-22-6-6-6-15-.01.

of cut constant at 0.031 inch, while the feed was varied from 0.006 to 0.030 inch; and, second, by maintaining a cutting feed of 0.0151 inch per revolution, while the depth of cut was varied from 0.005 to 0.100 inch. The cutting speed was kept constant at 25 feet per minute.

The results of the tests are shown plotted on log-log paper in Fig. 5, with the force values plotted for constant feed over variable depth on the left, and for constant depth over variable feed at the right. Identical forces were obtained for the oil-quenched and the three quenched and drawn test logs. The values are indicated by the solid lines in Fig. 5, the highest of the group. The slope of the

[2]O. W. Boston and C. E. Kraus, "Machinability Measured by a Simple Tool," *Metal Progress*, September, 1932, p. 29.

constant feed line is 0.76, while the slope of the constant depth line is 0.97. This gives a resulting equation for these four steels (Nos. 2, 3, 4, and 5), Table I, of $F = 131{,}000\ d^{0.76}\ f^{0.97}$. Force values for the cold-finished bar are shown as short dashed lines just below the solid lines in Fig. 5, while those for the annealed bar are shown as long dashed lines. The resulting equations are shown with the curve. These curves indicate in general that the harder the material, the higher the cutting force, also, that the annealed bar gives the lowest cutting force. It is interesting, however, that the cutting force values for the quenched and the quenched and drawn bars are identical, whereas the cutting speeds for these bars, shown in Fig. 4, are raised with the higher drawing temperatures.

The slopes of both the constant depth and constant feed curves are less than 45 degrees, as indicated by the fact that the value of the exponent of $d$ and $f$ in the force equations are less than unity. This indicates that the more efficient cutting, from a power standpoint, can be done by taking cuts with the heaviest feed and depth possible. In this case, a greater depth of cut is of more importance than large feed because of the lower exponent of the depth.

## Surface Quality

The surface quality resulting from turning log 3, quenched at 1520 degrees Fahr. and drawn at 600 degrees Fahr., was studied. Using a depth of cut of 0.10 inch and a feed of 0.0127 inch with a Stellite tool having a shape of 8-14-6-6-6-15-$\frac{3}{64}$, the log was turned for short distances at each of several cutting speeds on a diameter of 2.8 inches. The surfaces as obtained on the cylindrical log are shown in the photograph at the top of Fig. 6. The cutting speed in feet per minute appears just below each banded ring. Twelve close-up views of each surface are shown for each of the speeds from 20 to 220 feet per minute. These photographs were taken with the illumination of a 100-watt lamp shining through tracing paper.

The surface at 20 feet per minute is poor, at 40 feet per minute poorer, and at 60 feet per minute it is about the worst of all. An improvement is shown at 80 feet per minute, which continues until the maximum speed of 220 feet per minute is obtained (upper half of the photograph), at which time it appears that the built-up edge on the face of the tool has receded from the cutting edge and the cutting edge actually is removing the metal, producing a smooth

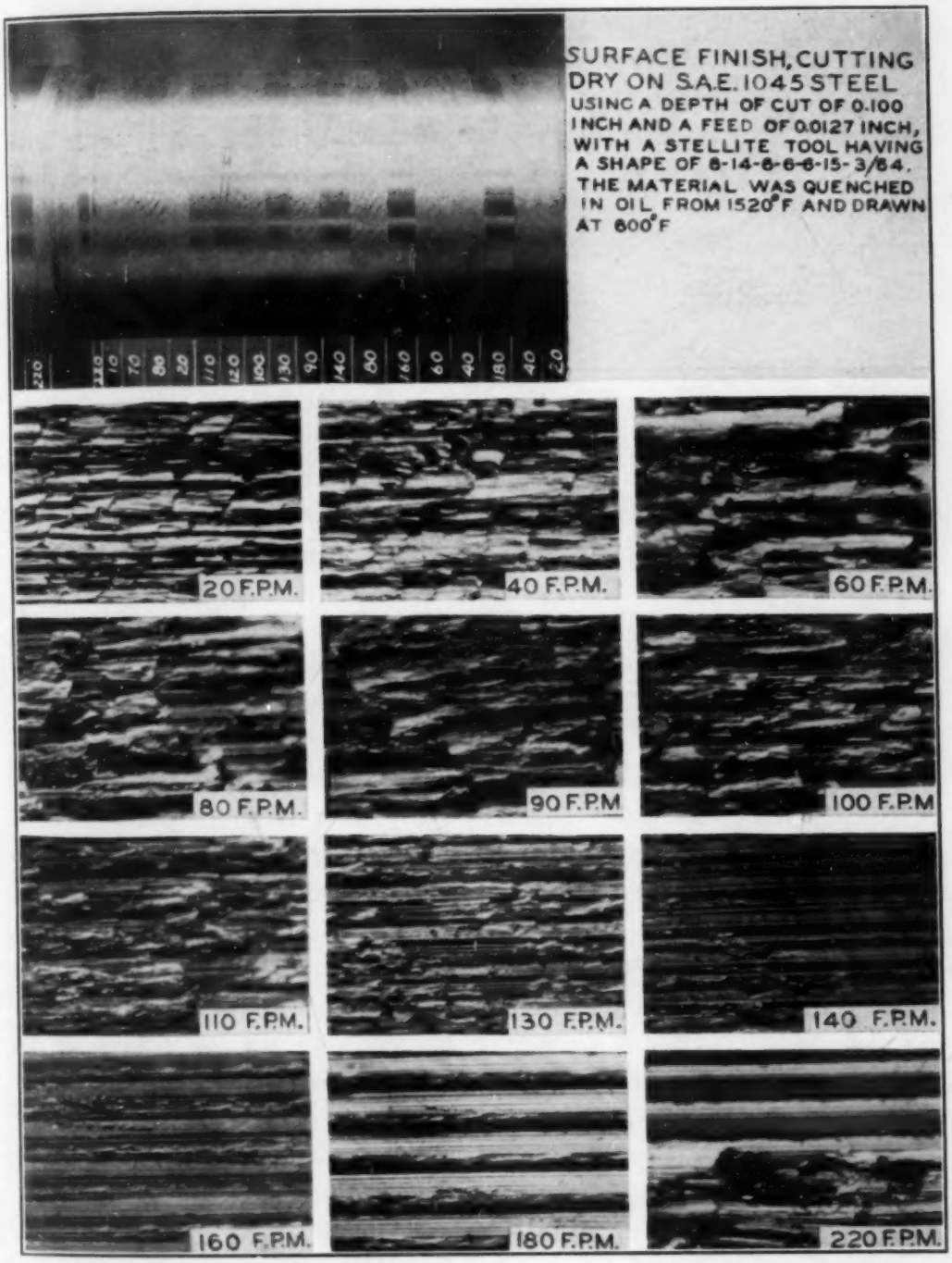
SURFACE FINISH, CUTTING DRY ON S.A.E. 1045 STEEL USING A DEPTH OF CUT OF 0.100 INCH AND A FEED OF 0.0127 INCH, WITH A STELLITE TOOL HAVING A SHAPE OF 8-14-6-6-6-15-3/64. THE MATERIAL WAS QUENCHED IN OIL FROM 1520°F AND DRAWN AT 600°F
20 F.P.M.
40 F.P.M.
60 F.P.M.
80 F.P.M.
90 F.P.M
100 F.P.M
110 F.P.M.
130 F.P.M.
140 F.P.M.
160 F.P.M.
180 F.P.M.
220 F.P.M.

Fig. 6—Surface Finish, Cutting Dry an S.A.E. 1045 Steel, Using a Depth of Cut of 0.100 Inch and a Feed of 0.0127 Inch, With a Stellite Tool Having a Shape of 8-14-6-6-6-15-$\frac{3}{64}$. The Cutting Speed Was Varied. The Material Was Quenched in Oil From 1520 Degrees Fahr. and Drawn at 600 Degrees Fahr.

highly burnished or clean-cut surface. The feed lines appear to be the most objectionable part to this finish. These pictures show conclusively that the surface quality varies a great deal with a change in speed, and in this case it is best at the highest speed obtainable.

It is poorest, not at the lowest speed, but at about 60 feet per minute.

In another series of tests, surface quality was studied for all six logs. A disk 3 inches in diameter and ½ inch thick was cut off from each log. These disks were, in turn, chucked in a lathe and faced to produce a standard surface prior to the final operation. The disk then was rotated at 312 revolutions per minute. A Stellite tool having a shape of 8-14-6-6-6-15-$\frac{3}{64}$, with the shank clamped at right angles to the face of the disk, was fed from the center to the outside at a uniform rate. The feed was 0.0085 inch per revolution, and the depth of cut was 0.005 inch. The cutting speed was 245 feet per minute maximum at the 3-inch diameter; 164 feet per minute at the 2-inch diameter; and 82 feet per minute at the 1-inch diameter. After the facing operation, a profilograph record was obtained for each log, as shown in Fig. 7. The profilograph needle traveled from A at the outside periphery to the point B adjoining the center hole, as shown at the top in Fig. 7. The record, similarly, goes from A at the right to B at the left. The metal is on the under side of the line. The vertical magnification of the surface finish was 250, while the horizontal was only 3.75. The scale may be observed from the 3-inch diameter disk.

The profilogram for the cold-finished bar, shown at the top, is uniformly irregular from A to B. The line is slightly concave at the center, indicating that a large built-up edge caused the machined surface to be formed at a greater distance from the cutting edge of the tool.

For the oil-quenched specimen, second from the top in Fig. 7, a bright burnished finish was obtained for about ¼ inch inward from the outer edge. This is shown on the profilogram for this disk, as the irregularity is due principally to the feed marks, inasmuch as the needle of the profilograph was drawn at right angles across the feed marks. It is significant, however, that the sides of the lines joining the feed marks are quite smooth. As the speed drops below 200 feet per minute, a built-up edge is formed on the tool face which produces a rough surface on the balance of the face of the disk.

In the quenched and drawn at 600 degrees Fahr. disk, the surface quality is seen to be superior from the outside to a depth radially of about ⅜ inch. In this case, even the feed marks are not so pronounced as for the oil-quenched specimen above. For all cutting speeds below 175 feet per minute, the surface quality is poor. The surface quality of the log quenched and drawn at 800 degrees Fahr.

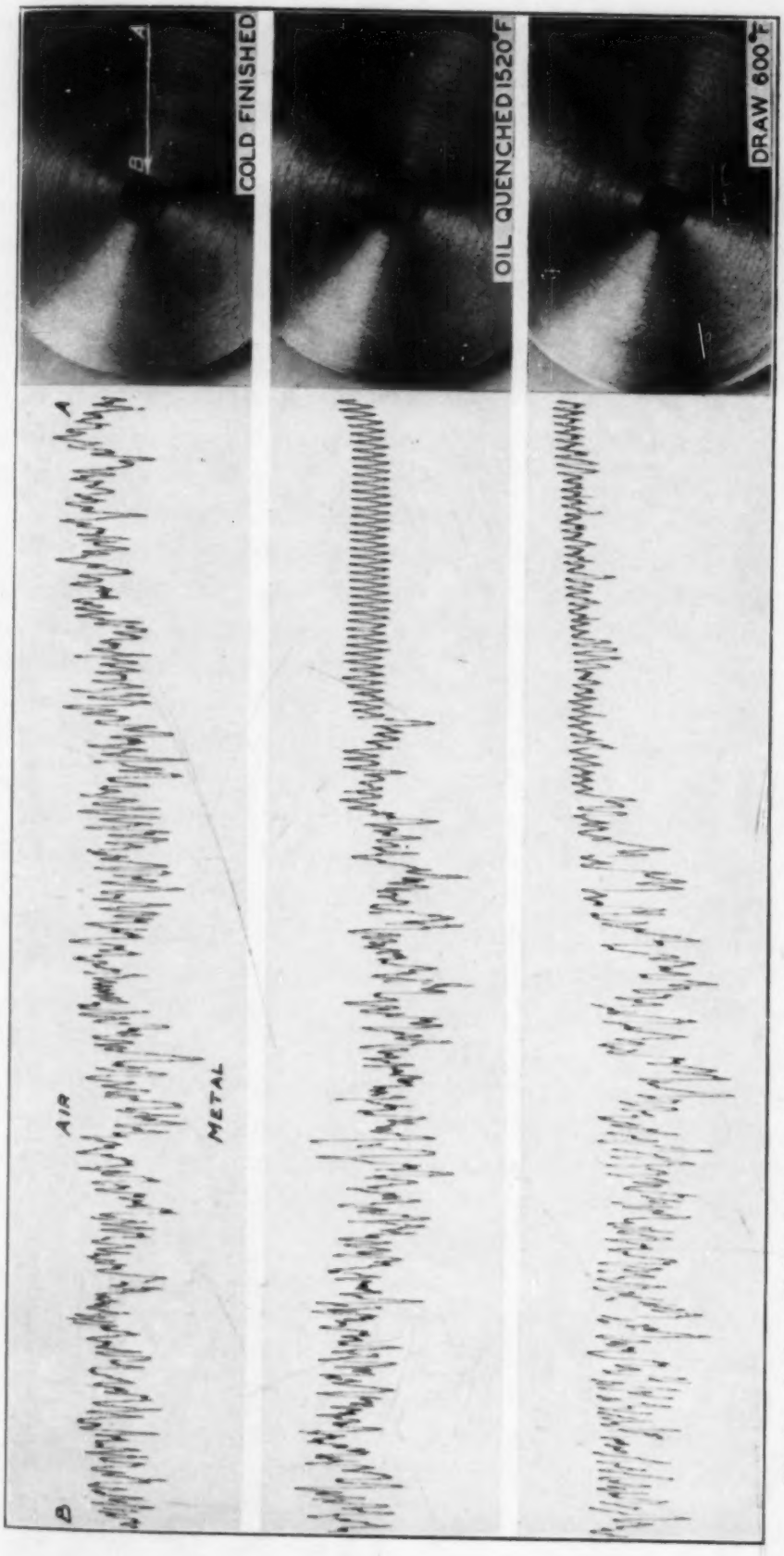


Fig. 7a—Surface Finish of S.A.E. 1045 Steel. 312 rpm. Maximum Cutting Speed 245 fpm. Depth 0.005 Inch, Feed 0.0085 Inch. Profilogram Vertical × 250. Horizontal × 3.75. Photograph × 0.5.

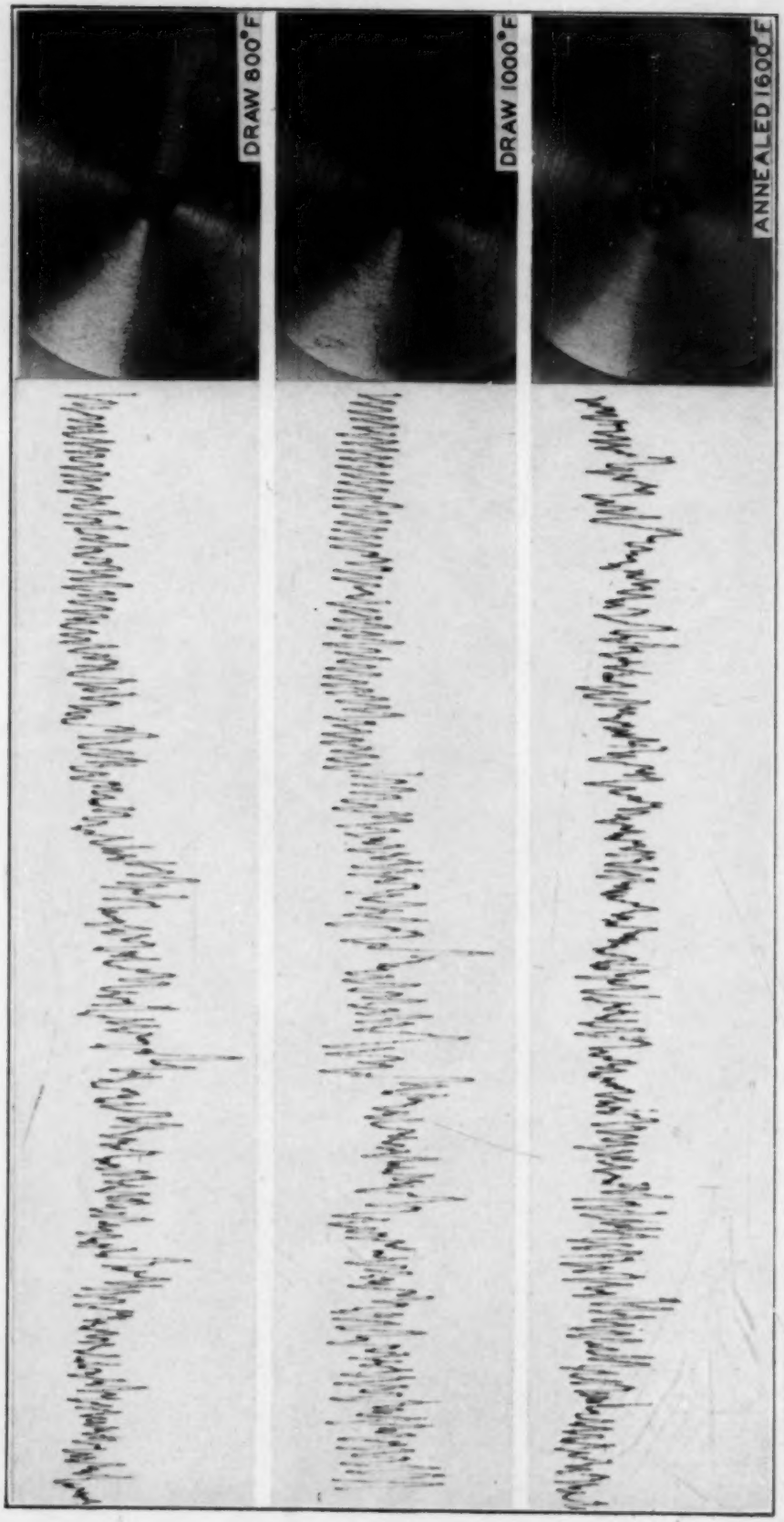


Fig. 7b—Surface Finish of S.A.E. 1045 Steel. 312 rpm. Maximum Cutting Speed 245 fpm. Depth 0.005 Inch, Feed 0.0085 Inch. Profilogram Vertical × 250. Horizontal × 3.75. Photograph × 0.5.

appears to be better at the outside diameter than toward the center. It is not so smooth as the oil-quenched disk and the disk quenched and drawn at 600 degrees Fahr. For the 1000 degrees Fahr. draw, the surface quality is poor, being even rougher than that for the cold-finished bar. The disk given the annealing treatment, shown at the bottom of Fig. 7b, also has a poor surface quality over its whole face. This study shows that the surface quality is definitely improved by the hardening treatment with or without the low temperature draw. The best finish occurs in all cases at the highest speed, although for the cold-finished and the annealed disks, the influence of speed on surface quality is not evident.

Figs. 8, 9, and 10 show macrographs of the finished surfaces at 3 inches diameter, 2 inches diameter, and 1 inch diameter, at which the surface cutting speeds are approximately 245, 164, and 82 feet per minute. The profilogram at the left was taken with a magnification of 500 vertical and 30 horizontal. These profilograms show to better advantage the nature of the surface of the oil-quenched and the oil-quenched and drawn logs. The feed lines appear to account for the greatest irregularity. The surface quality of the cold-finished and the annealed disks is shown to be irregular even at this high speed. The profilograms at the 2-inch diameter, where the speed is 164 feet per minute, as shown in Fig. 9, again show the surface quality to be poor. Of the group, that for the annealed bar appears to be the best. At the lowest speed of 82 feet per minute, taken at the 1-inch diameter, the finishes are universally poor, although the best finish appears to be obtained for the material in the quenched and drawn at 800 degrees Fahr., or in the annealed condition. All finishes at this low speed are poor, however.

These finishes were made on the inner face of the cut-off disks. Hardness values were obtained for each log at several distances from the center as shown in Table II. In most cases, the hardness at the 1¼-inch diameter is slightly higher than at the dead center. However, the difference in hardness from the 1-inch diameter to the 2½-inch diameter is not so great.

## Correlation

An effort was made to establish a relation between the physical and cutting properties of the various logs. Data are shown plotted on cartesian co-ordinates in Fig. 11. The logs are arranged from left

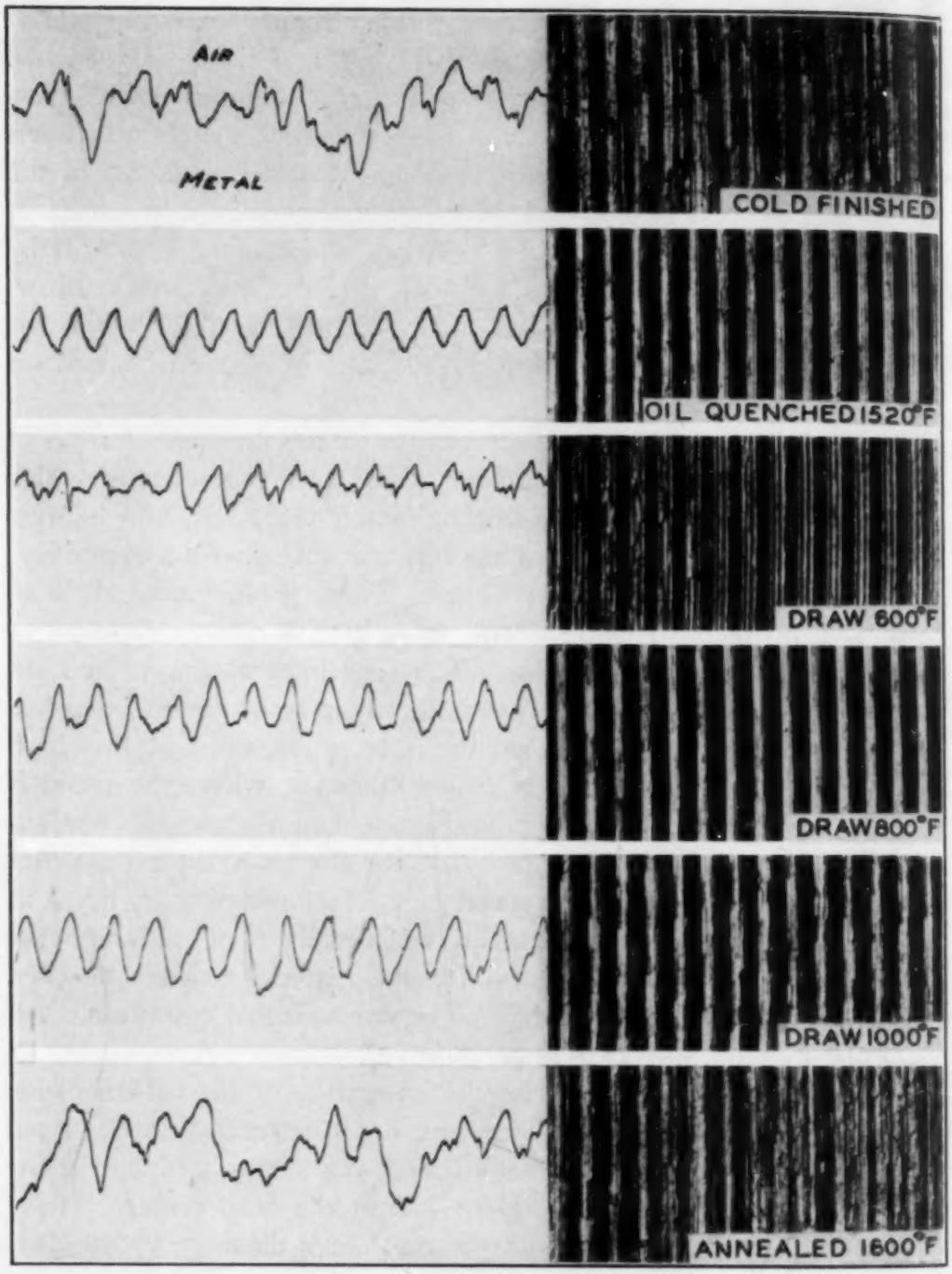


Fig. 8—Surface Finish of S.A.E. 1045 Steel. Maximum Cutting Speed 245 fpm. Depth 0.005 Inch, Feed 0.0085 Inch. Profilogram Vertical × 500. Horizontal × 30. Micrograph × 21.

to right in order of reducing ultimate strength and increasing cutting speed for a 1-minute tool life, $V_1$. This shows that the cutting speed increases as the ultimate strength decreases. The yield point, except for the cold-finished log 1, is practically equal for all logs.

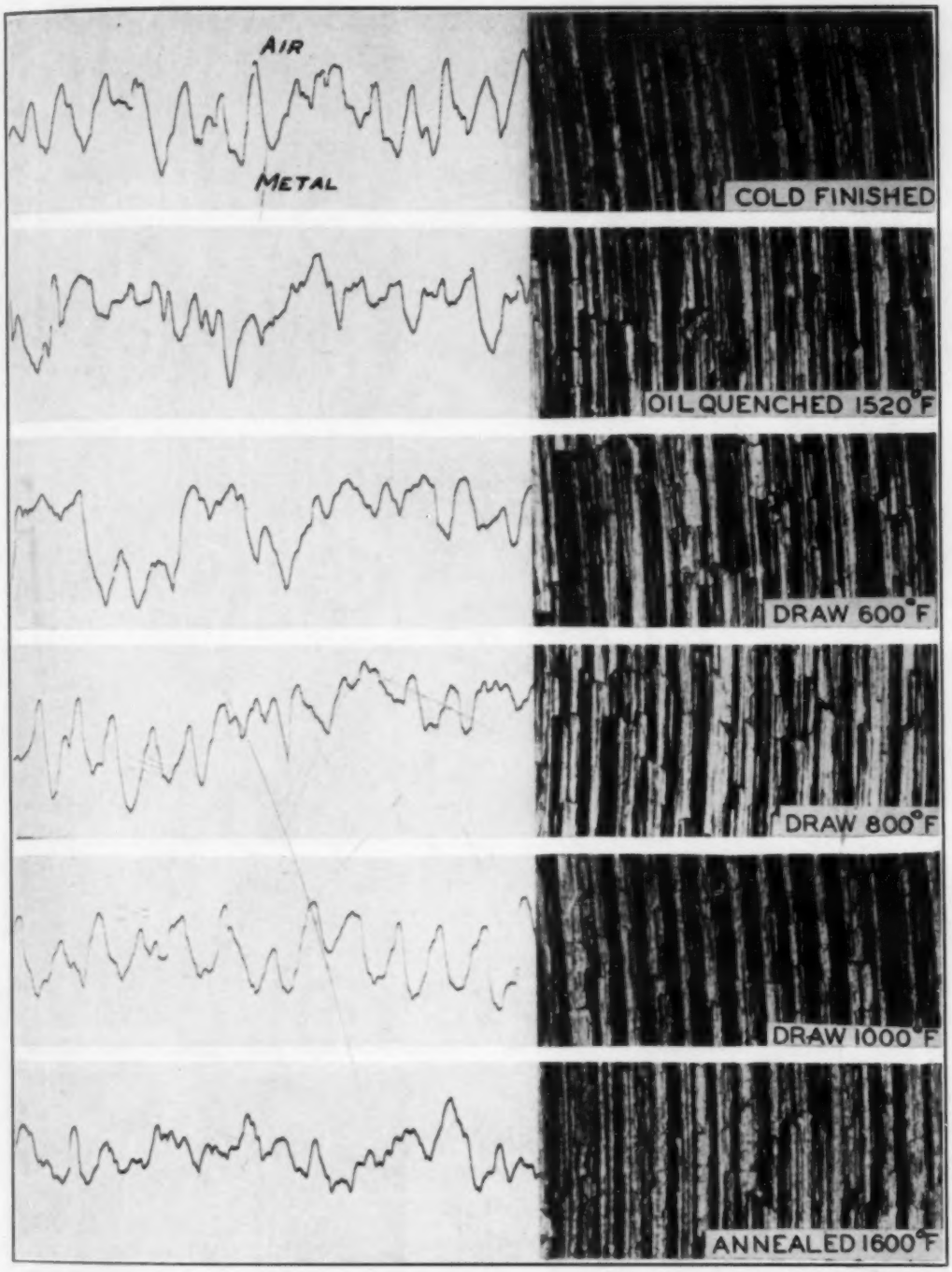


Fig. 9—Surface Finish of S.A.E. 1045 Steel. Maximum Cutting Speed 164 fpm. Depth 0.005 Inch, Feed 0.0085 Inch. Profilogram Vertical × 500. Horizontal × 30. Micrograph × 21.

The reduction of area, except for the log drawn at 1000 degrees Fahr., is reduced as the ultimate strength is reduced. The tangential cutting force and the Rockwell "B" hardness values taken on the logs at a diameter of 2½ inches also are shown to fall off with the

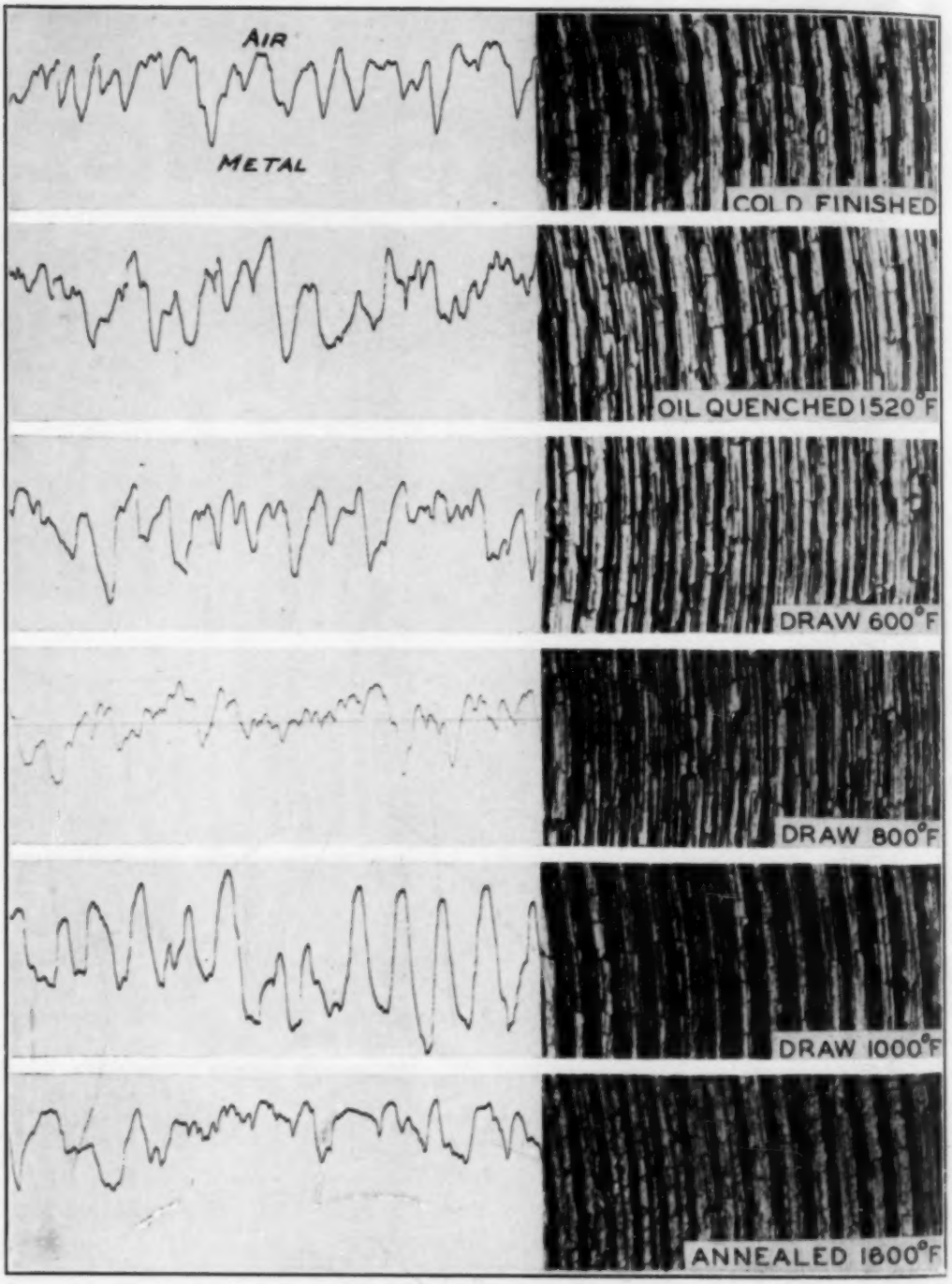


Fig. 10—Surface Finish of S.A.E. 1045 Steel. Maximum Cutting Speed 82 fpm. Depth 0.005 Inch, Feed 0.0085 Inch. Profilogram Vertical × 500. Horizontal × 30. Micrograph × 21.

ultimate strength. The elongation, except for the cold-finished bar 1, increases as the ultimate strength is reduced. The cutting temperature, as obtained with the thermocouple tool having the point formed of a tungsten carbide bit on the left and a Stellite bit on the right,

**Table II**

**Rockwell "B" Hardness on S.A.E. 1045 Steel 3-inch Diameter Disks Cut Off of End of Log (a)**

| Distance from Center | Rockwell "B" Hardness | | | | | |
|---|---|---|---|---|---|---|
| | Log 1 | Log 2 | Log 3 | Log 4 | Log 5 | Log 6 |
| 0 | 92.0 | 93.1 | 93.4 | 94.4 | 91.6 | 78.5 |
| ¼ | 92.6 | 93.2 | 95.5 | 94.4 | 93.0 | 79.3 |
| ½ | 93.2 | 96.0 | 96.8 | 96.4 | 94.1 | 79.9 |
| ¾ | 94.8 | 97.5 | 99.1 | 97.7 | 95.6 | 80.5 |
| 1 | 96.5 | 100 | 100.7 | 99.5 | 96.9 | 82.2 |
| 1¼ | 96.8 | 102.3 | 103.0 | 101.0 | 97.9 | 82.7 |

(a) For heat treatment of each log, See Table I.

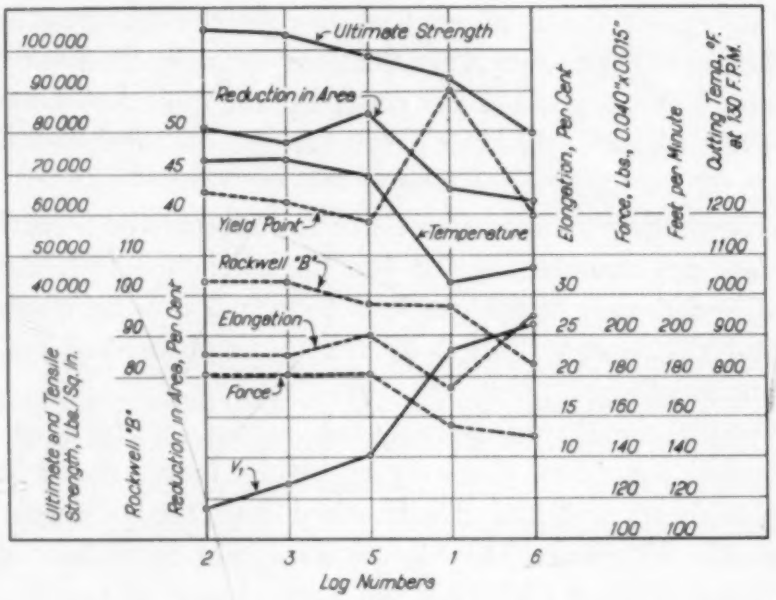


Fig. 11—Summary Curves With Various Factors Plotted Over Log Number Arranged, Left to Right, in Order of Increasing $V_1$.

when turning each log at 150 feet per minute with a depth of cut of 0.035 inch and a feed of 0.010 inch per revolution is shown plotted. The tool had a back rake of 6 degrees and a side rake of 6 degrees, and a nose radius of $\frac{1}{16}$ inch. These temperatures are shown for the heat treated logs to be 1320 degrees Fahr. for the quenched log: 1325 degrees Fahr. for the quenched and drawn at 600 degrees Fahr.; and 1285 degrees Fahr. for the quenched and drawn at 1000 degrees Fahr., showing a slight decrease as the drawing temperature is increased. The temperature for log 1 in the cold-finished condition is 1025 degrees Fahr., and for the annealed log 6 is 1060 degrees Fahr. The heat treated logs produce temperatures around 1300

degrees Fahr. as against 1050 for the cold-finished and annealed logs. In some respects, the temperature curve is the reverse of the cutting speed curve for a tool life of 1 minute, as might be expected.

## Conclusions

1. When turning variously treated logs of S.A.E. 1045 steel with high speed steel tools under standardized conditions as outlined, a relation between cutting speed and tool life for each log is obtained which gives a straight line on log-log paper corresponding to the equation $VT^n = C$. For these logs the lines are parallel. If the cutting speed for a given tool life of the oil-quenched log is represented as 100 per cent, then that for the quenched and drawn at 600 degrees Fahr. log is 111 per cent; for the quenched and drawn at 1000 degrees Fahr. is 124 per cent; for the cold-finished log is 168 per cent; and for the annealed log is 180 per cent.

Corresponding values of tool life in minutes for a cutting speed of 130 feet per minute are 0.3 for the quenched log; 0.8 for the quenched and drawn at 600 degrees Fahr.; 2 for the quenched and drawn at 1000 degrees Fahr.; 37 for the cold-finished log; and 64 for the annealed log.

2. The tangential turning forces are higher for the harder metals. The annealed log gives the lowest cutting force, the cold-finished log next highest, and the oil-quenched and the oil-quenched and drawn at 600, 800, and 1000 degrees Fahr. are equal but the highest values. More efficient cutting, from a standpoint of power per cubic inch of metal removed, can be done by taking the heaviest possible cuts.

3. When turning the quenched and drawn at 600 degrees Fahr. log at various speeds, the quality of the machined surface is improved as the cutting speed is increased. The best surface quality is obtained at the highest speed.

4. In facing tests on disks cut from each log, in which the cutting speed is increased uniformly from the center to a maximum at the outer diameter, the best finish is obtained at the highest speeds for the oil-quenched log. The next best finish is obtained at the highest speed for the quenched and drawn at 600 degrees Fahr. log, and similarly the third best finish is obtained from the oil-quenched and drawn at 800 degrees Fahr. log at the highest speed. A poor finish

is obtained for all bars at the lower speed ranges. At the highest speed range of 245 feet per minute, the poorest finish is obtained for the cold-finished and annealed logs. For the intermediate speed of 164 feet per minute, the best surface quality is obtained for the quenched and drawn at 1000 degrees Fahr. and the annealed logs, but all surface quality is poor at this speed. At the lowest speed of 82 feet per minute, the best surface quality is obtained for the log quenched and drawn at 800 degrees Fahr., with that for the annealed log a close second.

5. The cutting speed for a specific tool life is shown to increase for each log as the ultimate strength of the log is reduced. The hardness and the tangential cutting force are shown to fall off as the ultimate strength is reduced.

6. Cutting force cannot be used to predict tool life for quenched and drawn bars.

7. When operating at cutting speeds that would give the same tool life for all logs, the quenched logs and the quenched and drawn logs have a better surface finish than the annealed log.

### Acknowledgments

The steel logs used in these tests were submitted gratis through C. H. Beach, representative of Bliss and Laughlin, Inc.

The work was done in the Metal Processing Laboratories of the University of Michigan. Those helping to carry out the various tests are: D. J. Wangelin, L. P. Hartung, K. B. Kaiser, and Frank Freeman.

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## DISCUSSION

**Written Discussion:** By J. D. Armour, chief metallurgist, Union Drawn Steel Division, Republic Steel Corp., Massillon, Ohio.

All those interested in the subject of machinability will welcome this latest contribution of Boston and Gilbert to the fund of machinability data that has been increasing so rapidly during the past few years.

This paper brings out very clearly the fact that when we speak of machinability, we must state whether we refer to surface cutting speed, tool life, finish, or production. By production we mean the steel that will give the greatest number of satisfactory parts at the lowest machining cost. The authors concluded in this experiment that the quenched logs and the quenched and drawn logs have a better finish than the annealed log. This conclusion might be interpreted to mean that the quenched logs and the quenched and drawn logs showed better machinability than the annealed log. However, on

page 192, the authors show that at a cutting speed of 130 feet per minute, the tool life of the oil-quenched log was only 0.3 of a minute, while the tool life of the cold drawn log was 37 minutes and that of the annealed log 64 minutes. In other words, the cold drawn log at this cutting speed showed 12,200 per cent increased tool life and the annealed log showed 21,200 per cent increased tool life. From the production standpoint, the annealed log would undoubtedly be considered to have the best machinability.

We believe it is unfortunate that the authors selected oil-quenched 3-inch round S.A.E. 1045 logs to compare with cold-finished and annealed logs in this experiment. S.A.E. 1045 in a 3-inch section is very shallow hardening when quenched in oil, and the hardness, therefore, will drop off quite rapidly from outside to center. It is evident from the photomicrographs on the 2½-inch diameter that even at ¼ inch below the surface the quenching speed was not sufficiently rapid to retain all of the ferrite in solution. We believe photomicrographs close to the surface would have shown the ferrite entirely in solution and the physical properties in this same area would have shown higher hardness and tensile strength and lower elongation and reduction of area. Since the depth of cut on the tool life test was only 0.10 inch, the entire tool life test was conducted on material on which there is no record of either hardness or tensile properties. We believe this explains why there is such a wide divergence in cutting speed for a tool life of 1 minute compared to the variation in physical properties.

Since the cold-finished and annealed logs did not have this objection, a direct comparison can be made between these two test logs, and we see that the annealed log showed a rating of 180 per cent against 168 per cent for the cold-finished log in cutting speed for tool life of 1 minute. The annealed log showed an ideal structure for machinability on this type of steel, that is, a clear-cut blocky structure with the carbon present as lamellar perlite, whereas the cold-finished log showed an envelope structure with carbon present largely as sorbite, which is a very poor machining structure. If an annealed log, subsequently cold drawn, had been included in the test, we believe it would have shown a still better rating than the annealed log.

The paper brings out very clearly that there is a critical cutting speed to obtain maximum finish for any given depth of cut and material. In the tests on the cold drawn and annealed logs, this critical cutting speed had not yet been attained.

The data obtained by the use of the profilogram is very interesting. On the oil-quenched log and the log drawn at 600 degrees Fahr. it indicates that the material near the outside of the log was harder and lower in ductility than the material farther from the surface. The profilogram on the oil-quenched log indicates that the depth of hardenability in oil quenching of this section was approximately ¼ inch.

**Written Discussion:** By James Sorenson, metallurgist, The Four Wheel Drive Auto Co., Clintonville, Wis.

The authors of this paper are to be congratulated upon the thoroughness and the excellence shown in the presentation of the machinability data on S.A.E. 1045 steel.

Of particular interest is the method used in measuring surface finish in which the authors have arrived at a very unique and practical method of comparing the difference in surface finish which is of utmost importance in evaluating relative machinability.

In paragraph 4 under conclusions, the authors state the best finish is obtained at the highest speed for the oil-quenched log while the next best finish is obtained at the highest speed for the quenched and drawn at 600 degrees Fahr. log. I have been wondering if the authors have any explanation why there should have been this noticeable difference in surface finish when there is very little if any difference in microstructure and hardness between the oil-quenched log and the quenched and drawn at 600 degrees Fahr. log.

Also, I note that the authors state more efficient cutting, from a standpoint of power per cubic inch of metal removal, can be done by taking the heaviest possible cuts, which statement coincides with our experience in metal removal tests which we recently conducted.

### Authors' Reply

The authors are indebted to Messrs. J. D. Armour and James Sorenson for their discussions of this paper.

Mr. Armour has pointed out that the quenched and the quenched and drawn logs have better finish than the annealed logs, whereas the annealed logs had the longer tool life. When discussing machinability, the finish and the tool life must be considered independently. Power required to cut the metal, also, may be used as a measure of machinability, but again is independent of either the tool life or surface finish. Therefore, when stating machinability ratings, these ratings must be qualified on the basis of tool life, surface finish, or tool force.

The tool-life tests were not limited to one pass on the outside diameter of the 3-inch section. However, the tool-life tests were corrected so that they would represent tests run on this outside diameter. Very little correction was necessary, since usually not more than three passes were made. Hardness tests were made at all diameters so that the tool-life data represented could be compared directly with the hardness at the outside diameter. The tensile tests were conducted on bars taken on the inside diameter and, therefore, as pointed out by Mr. Armour, would not represent exactly the physical properties of the material cut in the tool-life tests.

Mr. Armour also mentions that the cold-finished log shows "an envelope structure with carbon present largely as sorbite which is a very poor machining structure." This material was supplied to the authors in the annealed and drawn condition, and photomicrographs taken from near the outside of the bars showed a lamellar pearlitic structure. The annealed bars showed a spheroidized-lamellar structure which may account for the increase in tool life. The authors feel that a soft anneal usually is better for long tool life, whereas the annealed and drawn material may be better for surface finish and good chip formation.

The surface finish at the outside of the disks shown in Fig. 7a is due more to high cutting speed than to variation in hardness from the inside to

the outside of the bar. This may be verified by referring to Fig. 6, in which surface finish tests were made at variable speeds on one diameter of the quenched and drawn at 600 degrees Fahr. log. These tests show definitely that speeds above 160 feet per minute give superior surface finish.

It is gratifying to learn that the tests conducted by Mr. Sorenson agree with those of the authors in that heavy cuts are the most efficient from a power standpoint. It is true also that heavier cuts are more efficient from a tool-life standpoint, in that a larger amount of metal can be removed in the same amount of time as heavier feeds and depths are used along with a decrease in cutting speed.

The best finishes were obtained with the oil-quenched log, and the next best finish with the log quenched and drawn at 600 degrees Fahr. The superiority of the oil-quenched log was not too evident, however. The photomicrographs do show a greater percentage of troostite which may account for the variation in tool life, as well as for the variation in surface finish.

# TIME QUENCHING

By John L. Burns and Victor Brown

*Abstract*

*Experiments have been conducted in order to study qualitatively the effects produced when the cooling rate of a piece of steel is interrupted during quenching. It is revealed that concentric rings, which differ structurally, is a typical condition produced. An explanation is offered as to the reasons for their occurrence on the basis of transformation shown by the "S" curve. Hardness distribution curves reveal some interesting and significant results and lend support to other points brought out by microscopic study. An analysis is given which may enable those who utilize this manner of heat treating in one form or another to obtain the best results.*

TIME quenching has been practiced by some heat treaters for many years, but in comparison with ordinary quenching its use is quite limited. For the most part it has been used principally as a means of compromising between straight oil and straight water quenching in order to obtain greater response than oil alone offers, yet without exposing a given part to the danger of cracking that straight water quenching entails. Utilization of its principle is found to a degree in heat treatment of tools and dies, yet if all its modified forms are considered it is used in general industry to an appreciable degree. Although results of some types of time quenching have been published, we find little data of an analytical character.

Since the term "time quenching" is a broad one, it may be well to consider some of its implications. First, the word "time" denotes that the period of exposure to a given quenching medium is a measured one. Beyond this the term has no other meaning. The simplest form of "time quenching" would be to expose an article to one type liquid quenching medium for a predetermined length of time, after which it is removed to the influence of the atmosphere. Other forms of "time quenching" would be "double quenching," exposing an article successively to two different types of liquid quenching media, at least one of which would be measured in the

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course of cooling to room temperature, and "interrupted quenching," the use of one liquid medium from which the article is withdrawn and then replaced in the course of cooling to room temperature; the time of quenching before and during the interruption is usually measured. It is apparent that other means of "time quenching" are possible, but the ones mentioned above are probably the most important.

An attempt is made to analyze the procedure of transformation during the stages of the three typical types of time quenching and to show effects in terms of microstructures and hardness distribution charts. S.A.E. 1040 and 1060 steels were used—the simple time quench was handled in water, the double quench consisted of first immersing in water with a quick change to oil, whereas the interrupted quench was carried out in water.

## Materials

Three different heats of fine-grained S.A.E. 1040 steel and three heats of fine-grained S.A.E. 1060 steel were used, some of which had a slight addition of chromium. Analyses of the various heats are listed in Table I.

The general procedure consisted of forging 1¾-inch round bars from billets after which they were normalized at 1650 degrees Fahr. (900 degrees Cent.). Samples 1½ inches round and 4 inches long were then machined from the forgings.

## Simple Time Quenching

Samples from heats 1A and 1B, used in this series of tests, were held at 1500 degrees Fahr. (815 degrees Cent.) for 45 minutes and quenched as follows:

(1) Air-cooled only
(2) 5 seconds in water—finish cooling in air
(3) 10 seconds in water—finish cooling in air
(4) 15 seconds in water—finish cooling in air
(5) Water-quenched (complete).

Each sample was cut transversely, through a section equidistant from both ends which was ground flat and subsequently polished and etched. Etching was carried out electrolytically in 15 per cent alcoholic HCl, as it gave an excellent contrast between the different constituents produced—usually in concentric rings. After recording

**Table I**
**Composition of Steels Used**

| Heat | C | Mn | P | S | Si | Cr | Grain Size |
|---|---|---|---|---|---|---|---|
| 1040 | | | | | | | |
| 1A | 0.41 | 0.77 | 0.020 | 0.022 | 0.234 | .... | 6/7 |
| 2A | 0.44 | 0.80 | 0.022 | 0.028 | 0.272 | .... | 6/7 |
| 3A | 0.44 | 0.70 | 0.018 | 0.033 | 0.250 | .... | 6 |
| 1060 | | | | | | | |
| 1B | 0.60 | 0.81 | 0.029 | 0.035 | 0.269 | .... | 6/7 |
| 2B | 0.61 | 0.83 | 0.025 | 0.027 | 0.280 | 0.18 | 6/7 |
| 3B | 0.63 | 0.75 | 0.015 | 0.029 | 0.193 | 0.15 | 6 |

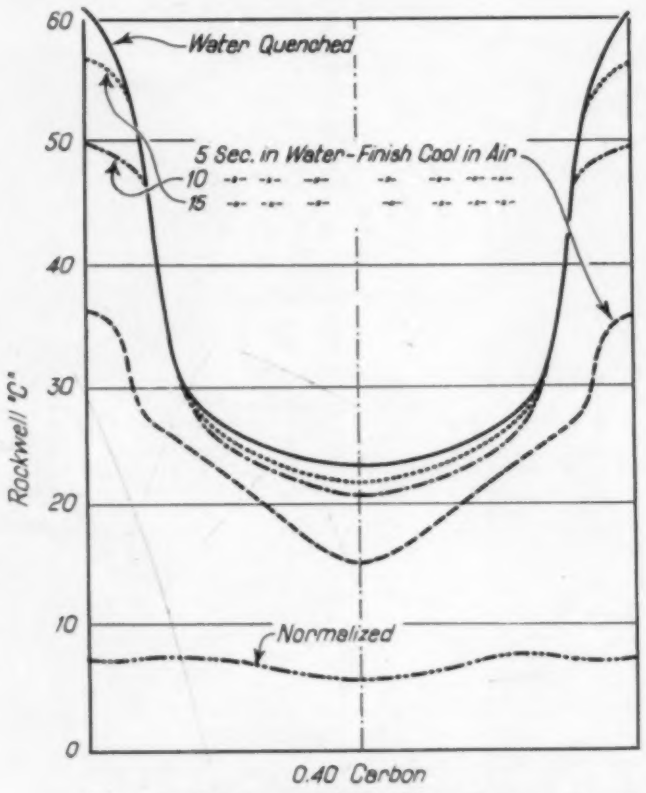


Fig. 1—Hardness Distribution Curves for the Simple Time Quench. Carbon Steel, 0.40 Per Cent Carbon.

the macroscopic appearance of the etched samples by photography, hardness distribution curves were produced in the usual manner by taking Rockwell impressions at measured distances from the surface along diameters drawn across the transverse face. All samples were handled in the same manner.

Hardness distribution curves for the simple time quench are shown in Fig. 1 (0.40 per cent carbon) and Fig. 2 (0.60 per cent carbon). The straight water quench and air-cooled tests are added for comparative purposes. The curves indicate a lowering of the

near surface hardness as the primary effect with the hardness through the remainder of the sections well sustained even with only a 5-second dip in water.

The photomacrographs of etched cross sections reveal the usual quenched outer ring (Fig. 3) produced during water immersion (only the 0.60 per cent carbon steel was photographed since the 0.40 per cent carbon material showed rather indistinct circles due to its lower hardenability when subjected to this type quench). A 15-sec-

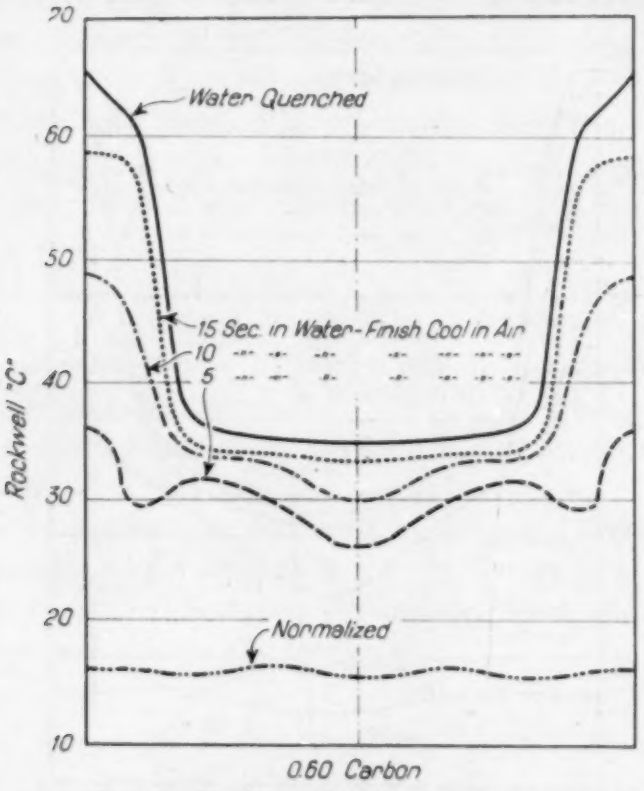


**Fig. 2—Hardness Distribution Curves for the Simple Time Quench. Carbon Steel, 0.60 Per Cent Carbon.**

ond water dip (sample No. 4) shows practically no difference from the straight water quench (sample No. 5), but a 10-second water dip (sample No. 3) reveals a distinct drawn martensitic ring at the surface. This is a result of drawing the quenched surface by the conduction of heat from the center of the bar at the time of its removal from the quenchant. Number 2 sample, quenched in water for 5 seconds, shows three distinct rings; structures at these locations have been photographed. The outer ring marked "A" reveals a fine quenched and drawn sorbite. The adjacent ring marked "B" contains a carbide-ferrite aggregate, some of which is primary

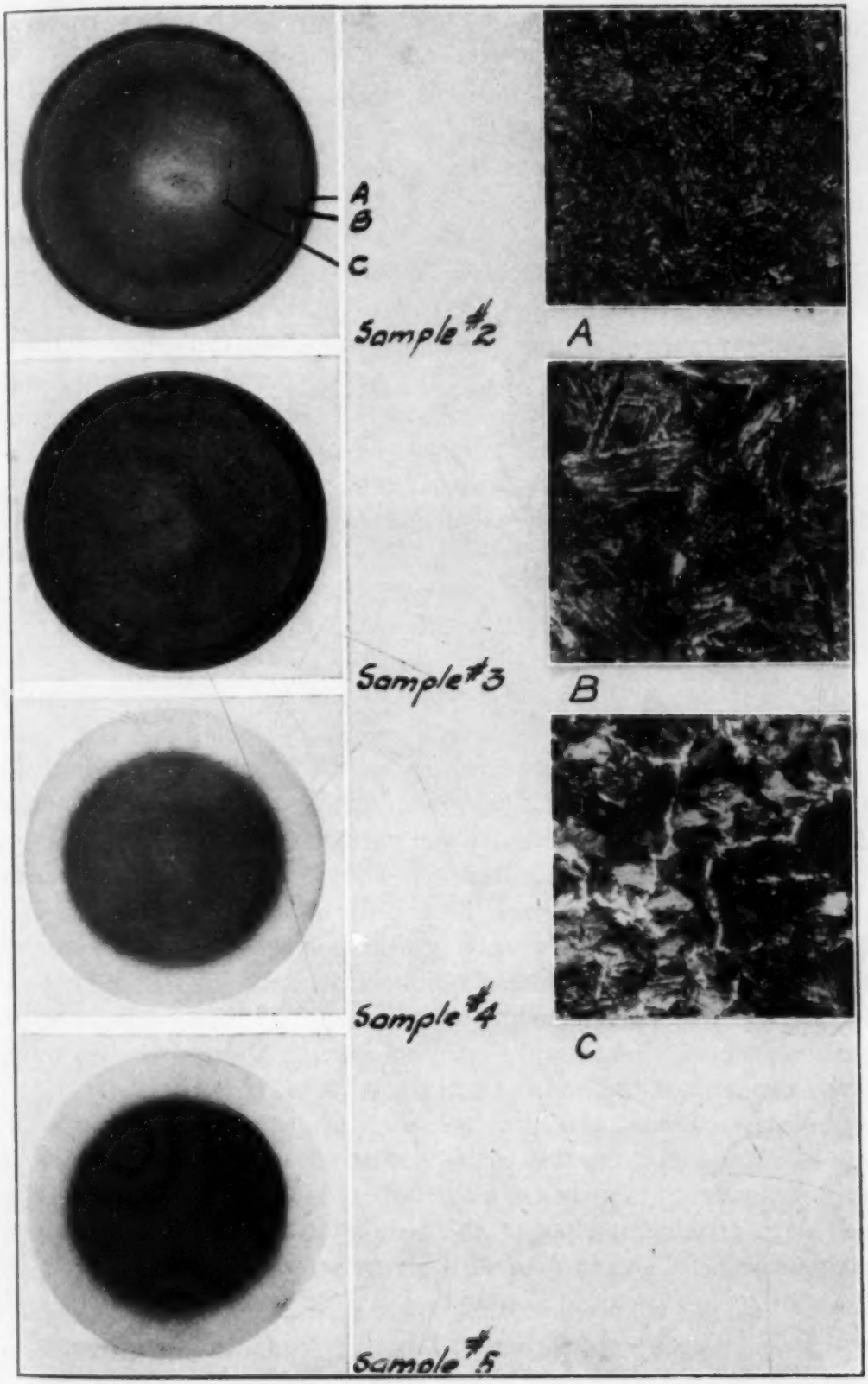


Fig. 3—Macrographs of Etched Cross Sections of 0.60 Per Cent Carbon Steel.

troostite while most of it is pseudo-martensite. The center is composed of sorbite plus ferrite network.

Mention has been made in the literature of rings obtained in steel cylinders which were subjected to time quenching[1] but they were not shown to the same degree nor has any attempt been made to analyze their occurrence.

The fluctuations in the hardness distribution curve of sample No. 2 (5 seconds in water) conform to and are the result of the different types of structures present as concentric rings in the sample. The presence of various structures indicates a difference of transformation in those rings. As in the next type of time quench with which this paper deals the transformations occurring at various points through the section can best be analyzed by using the "S" curve developed by Bain and associates.[2] The essential difference between this type of time quench and the others is that a general tempering or drawing takes place after finishing the predetermined period in the quenchant. Naturally substantial structural changes may then occur depending upon the timing.

It is significant to note that the combined depths of rings "A" and "B" of sample No. 2 is apparently greater than the quenched ring on the sample treated entirely in water (sample No. 5).

This manner of time quenching has been used quite extensively in regular production heat treatment of medium sized parts made from S.A.E. 1040, 1045 and water quenching alloy grades such as S.A.E. 5130. Where an oil quench would be inadequate but where critical design is a handicap for a conventional full water quench, because of the danger of quench cracking, it has been demonstrated that by the use of a proper time quench a solution to the problem is offered. The parts shown in Figs. 4 and 5 might be included in such a problem; they could have been given a short period in water since the desired finished hardness was 241-286 B.H.N.

Various forged and machined parts are treated by time quenching—included are tractor parts, industrial machinery parts, bolts (approximately 1½ inches in diameter) and hand tools. They are subjected to treatments that fit them individually, but the quenching arrangement follows in general the type under discussion.

On some parts it is a simple matter to apply production meth-

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[1]M. A. Grossmann, "Principles of Heat Treatment," published by American Society for Metals, 1935.

[2]Research Laboratory, United States Steel Corp., "The Process and Result of Austenite Transformation at Constant Temperatures," *Metals and Alloys*, January 1937.

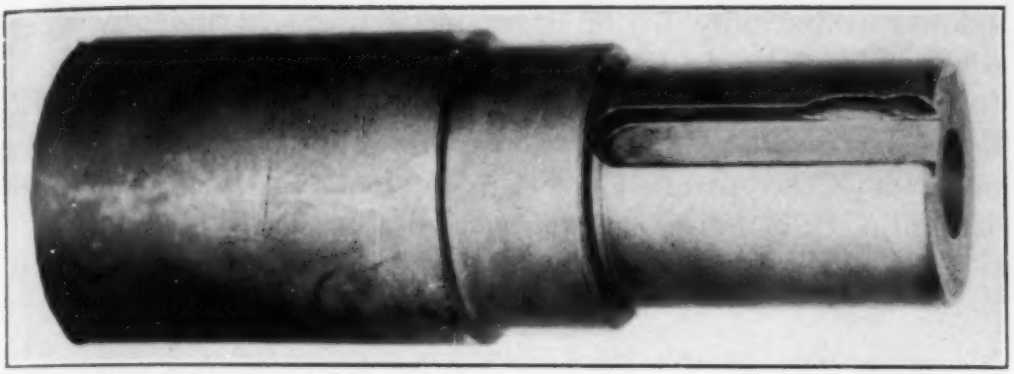

Fig. 4—Part Made of S.A.E. 1040 Steel, Water-Quenched. × ⅔.

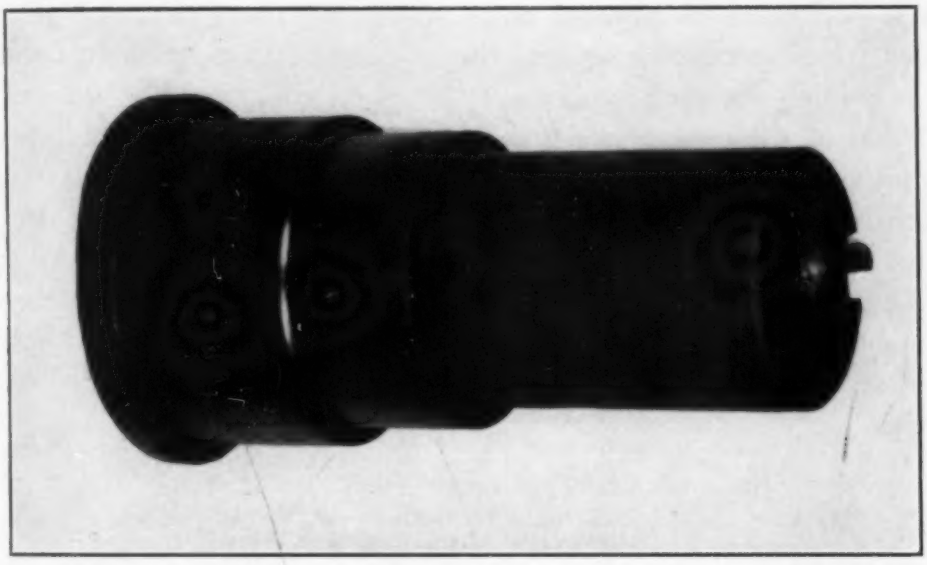

Fig. 5—Part Made of S.A.E. 1040 Steel, Water-Quenched. × ⅔.

ods to this manner of time quenching. Equipment consists merely of the usual heating furnace (it may be continuous) near which is placed a suitable water quenching tank. A conveyor, built into the tank, may be designed to carry the work out of the water automatically. It is readily apparent that timing is accomplished by adjusting the flow of material from the heating furnace into and out of the water. It is not inferred that with time quenching it is permissible to omit the drawing operation. With some parts it might be possible to do this, but only after careful consideration. Since parts come out of the quenchant in a heated condition, it requires only a slight extra cost to provide a thorough draw; it may be done continuously through a conveyer type drawing furnace. The temperature of the drawing furnace should always be equal to or in excess of the temperature of the work as it comes from the quenchant;

a lower temperature would prove ineffective. In the case where considerable variation in mass is involved a study of quenching effects, design and heat flow becomes necessary in order to arrive at the correct procedure and to obtain the properties where they are required. Other parts following more simple lines are more easily handled and thus cause comparatively less concern.

## Double Quenching

Heats 2A and 2B were used for this series of tests which consisted in heating samples to 1500 degrees Fahr. (815 degrees Cent.) for a period of 45 minutes then immersing them one at a time in water for a specified length of time, followed by a quick transfer to an adjoining oil tank where they finished cooling. The interval in air, that is, the time between the two quenching media, was approximately one second. The temperatures of the respective liquids were measured just prior to quenching, the oil being held to 120 degrees Fahr., while the water was maintained at 70 degrees Fahr. All samples entered the quenching media in a vertical position and were vigorously agitated for uniformity. The following time intervals were used on sets of samples from both the 0.40 and 0.60 per cent carbon steels.

| Sample No. | Quenching Conditions |
|---|---|
| 1 | Straight oil quench. |
| 2 | 2 seconds in water, transferred to oil. |
| 3 | 5 seconds in water, transferred to oil. |
| 4 | 8 seconds in water, transferred to oil. |
| 5 | 11 seconds in water, transferred to oil. |
| 6 | 14 seconds in water, transferred to oil. |
| 7 | Straight water quench. |

## Results

Both the 0.40 and 0.60 per cent carbon steels behaved similarly in this type of quench, showing concentric rings within the circumference. The effects produced were decidedly more pronounced in the 0.60 per cent carbon steel, due to its greater hardening response under the cooling conditions. The general effect under certain time limits was to produce a subsurface area, the extent of which was governed by the time factor, softer than otherwise obtained in straight water quenching. (Figs. 6 and 7). In a few cases, on the 0.60 per cent carbon steel, the actual surface hardness readings were lower than obtained by straight oil quenching, although the tendency

would then be to have a higher subsurface hardness, equaling or exceeding that of the oil-quenched surface hardness. Note samples No. 2 and No. 3 (0.60 per cent carbon grade), Fig. 7.

Within certain time limits, a dark outer ring appeared which microscopic examination revealed to be composed for the most part of tempered martensite. The width of this ring appeared greater as the time in water was increased up to a certain point. (Refer to Fig. 8 and hardness curves of Figs. 6 and 7). The contour of the

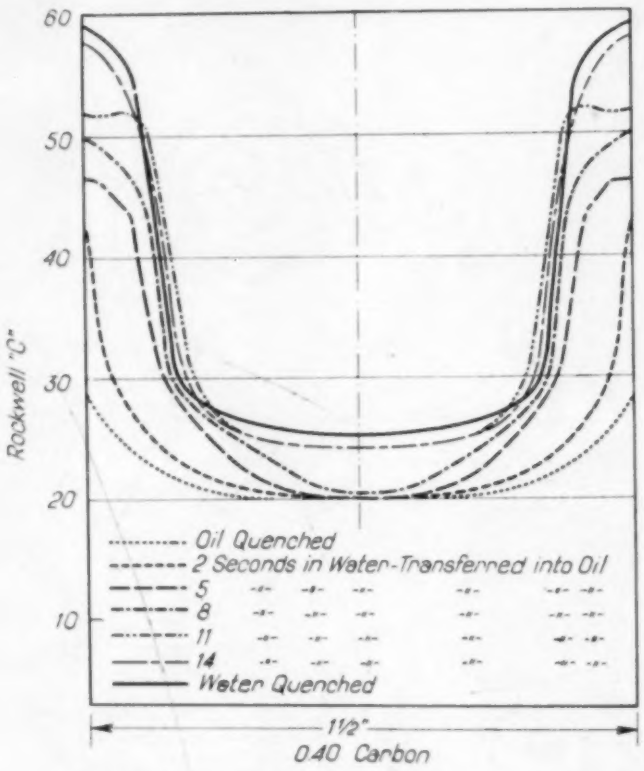


**Fig. 6—Hardness Distribution Curves for Oil-Quenched Carbon Steel. 0.40 Per Cent Carbon.**

hardness distribution curve, particularly the 0.60 per cent carbon grade, reflects this outer ring rather sharply. In the instance of sample No. 6 (Fig. 7), the curve definitely indicates the presence of such a ring, although it is not readily apparent after etching (Fig. 8). The white ring within the dark one is composed largely of martensite with a variable amount of tempered martensite, depending on the sample. The inner portion of the sample is composed chiefly of primary cooled structures, including areas of martensite, troostite, sorbite and ferrite, the last two constituents being confined to the center. Samples No. 3 and No. 4 seem to show four distinct areas, the

difference being in the inner dark zone, which is divided on these two samples into two parts. They are not reflected on the curves and are also entirely lacking on the 0.40 per cent carbon steel.

An explanation for the occurrence of various constituents in concentric rings can probably be most satisfactorily arrived at by referring to the "S" curve. On Fig. 9 there has been sketched a cool-

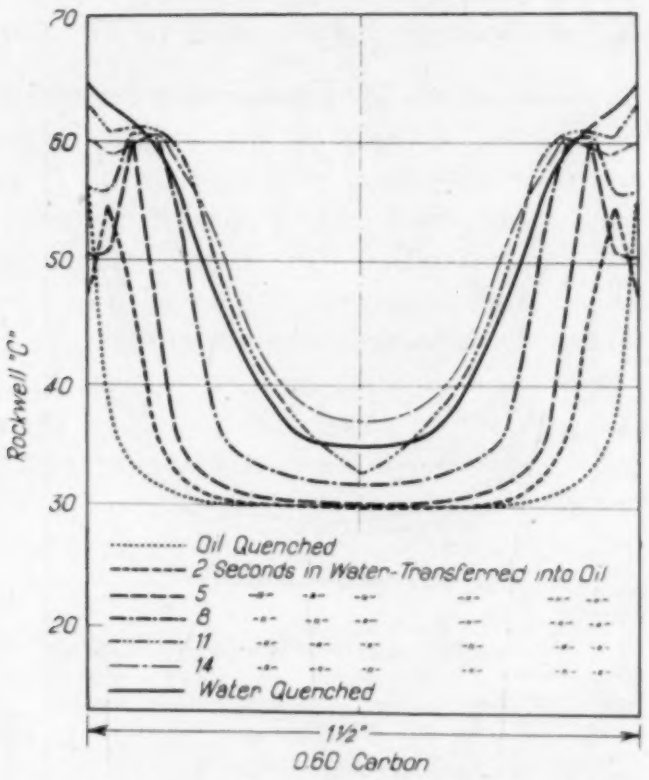


**Fig. 7—Hardness Distribution Curves for Oil-Quenched Carbon Steel. 0.60 Per Cent Carbon.**

ing curve of each zone or ring shown on sample No. 3, Fig. 8. These curves and their relationship with each other and also with a reproduction of the "S" curve are estimated but they are utilized in order to simplify explanation of the manner of transformation. The outer dark ring "A" apparently had been transformed to martensite during the initial quench (water dip) and during the interval in air became tempered by heat flowing outward from the center of the section. This condition is illustrated in Fig. 9 by a critical quench and the dotted line represents the drawing effect by heat flowing outward from the center. The white ring denoted by "B" reached an undercooled austenitic condition but failed to transform to martensite during the initial water quench. Instead, it remained in this condition

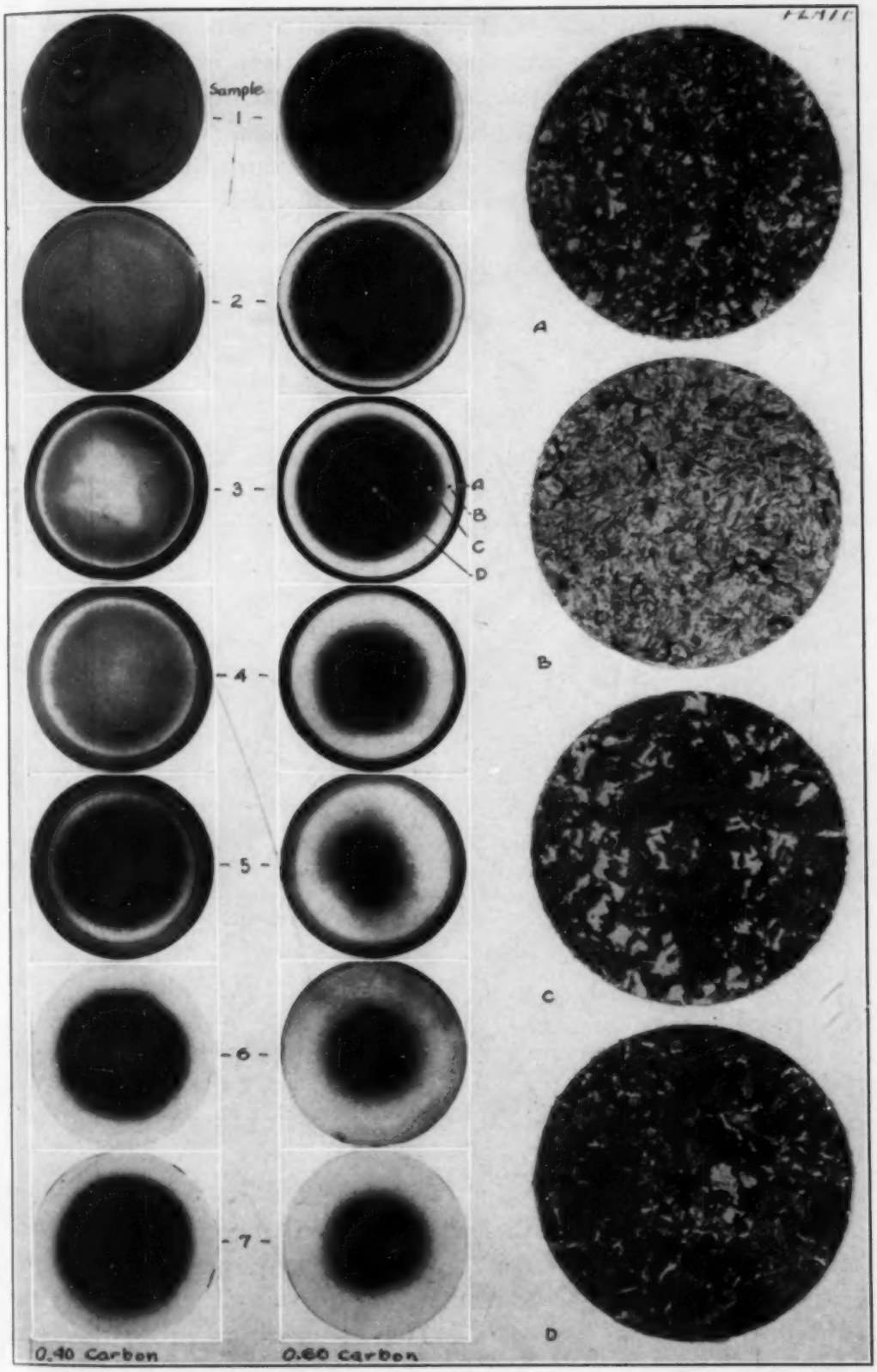


Fig. 8—Macrographs of Etched Cross Sections of 0.40 and 0.60 Per Cent Carbon Steels.

until the oil quench took effect and produced approximately a critical quench. Internal temperatures are progressively higher toward the center during the interval in air, therefore transformation here occurs in higher ranges. As shown by the cooling curves, ring "B" must be in a temperature range which corresponds to the lower portion of the "S" curve, where a comparatively greater time may elapse

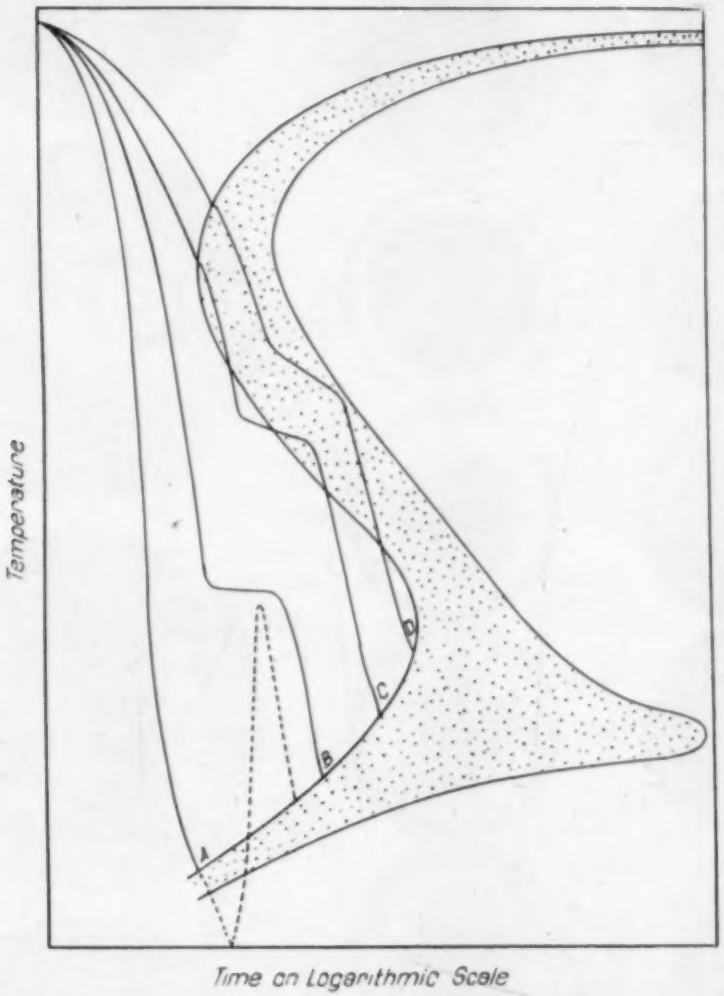


Fig. 9—S-Curve and Cooling Curve of Each Zone or Ring Shown on Sample 3, Fig. 8.

before transformation. Rings "C" and "D" are transformed at higher temperatures where earlier transformation occurs. As a result the structures produced will also differ. It may mean that portions corresponding to ring "C" become partially transformed during the interval in air, while there is a halt in temperature drop; this may actually take place prior to the transformation of ring "B," which is at a lower temperature, but where a longer period of time is afforded

in which to complete the transformation. Transformation is progressive and not instantaneous; thus producing transition zones as well as structural heterogeneity, as seen in the photomicrograph of ring "C."

It is found that with certain timing a hardness distribution curve can be made to fall inside the curve of an adjacent sample fully water-quenched. This may be due to the different type structure produced at portions represented by the particular part of the curve in question. In time quenching, a structure of pseudo-martensite is found dispersed throughout portions such as ring "C", sample No. 3, Fig. 8; could this increase the aggregate hardness? The fact that an outer quenched ring becomes immediately tempered will relieve quenching stresses. This in turn may have an effect on adjacent interior portions which have delayed transformation.

Generally we find that it is possible to depress the surface hardness in relation to subsurface hardness. This is true only on sizes and chemistry which would tend to produce a critical combination of sensitivity. It will be noticed that the comparatively low response to quenching of the 0.40 per cent carbon steel makes the rings, which are very apparent by etching, practically indistinguishable on the hardness curves. This is true since the hardness of the structural mixture of martensite and primary troostite in section "B" practically "balances" the hardness of the drawn outer martensitic ring (section "A") as measured with Rockwell tester. The 0.60 per cent carbon steel, due to its higher hardenability, offers a much better chance to note differences by both Rockwell hardness and structure methods when used in this size. By decreasing the size of the bars tested the contrary probably would be true, i.e., the 0.40 per cent carbon steel would exhibit the better combination of size and hardenability, whereas the 0.60 per cent carbon would quench out hard nearly all the way across the section under most of the testing conditions.

On comparing the macrophotographs of the 0.40 per cent carbon steel with the 0.60 per cent carbon steel, it is noted that generally the dark outer ring (section "A") is larger on the 0.40 per cent carbon steel. On the other hand, the white inner ring is wider on the 0.60 per cent carbon steel. This is undoubtedly due to the higher hardenability of the 0.60 per cent carbon steel.

The following is an example of this type of time quenching. Shear knives and certain die blocks made from S.A.E. 6150 steel

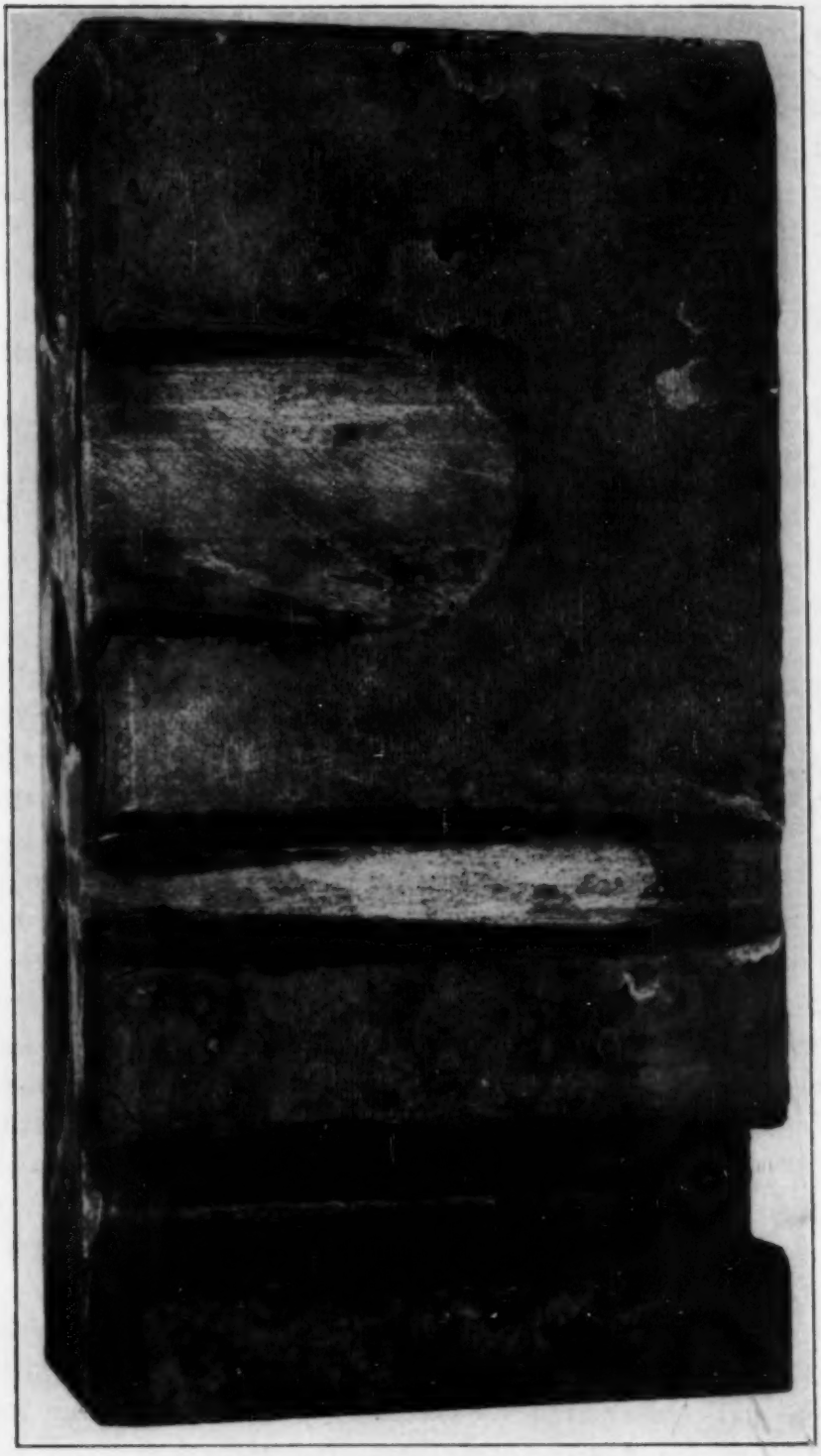

Fig. 10—Die Made of S.A.E. 6150 Steel. Water-Quenched. × ⅓.

had provided a problem to quench. Full water quenching had cracked some in the manner shown in Fig. 10 and it was found that oil quenching could not produce sufficient hardness due to the

mass involved. In order to fit them for service without further loss it was decided to time quench in water with a transfer to oil. This consisted in varying the time in water from 30 to 50 seconds (depending upon the size of the part), then transferring into oil. The results proved satisfactory and the treatment was adopted as standard procedure for quenching these parts.

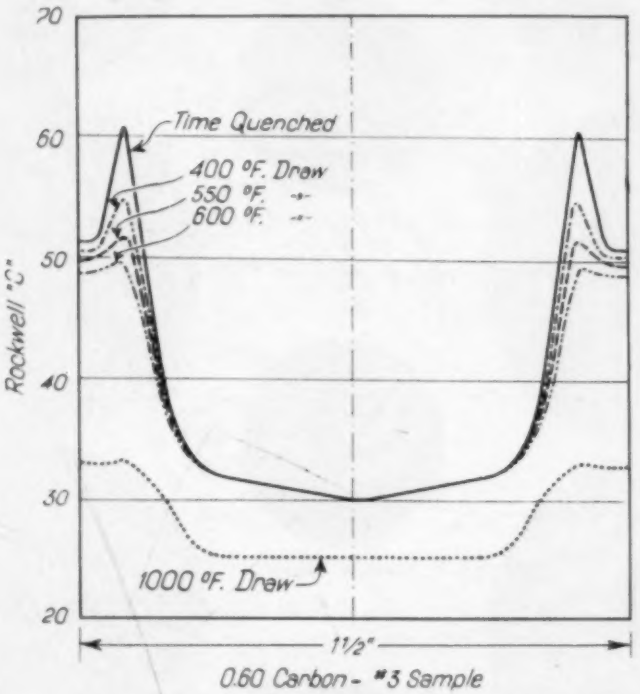


**Fig. 11—Hardness Distribution Curves for Time-Quenched and Drawn 0.60 Per Cent Carbon Steel.**

This merely serves to illustrate the possibilities for this method of quenching. It has been used for just such applications in the past and many heat treaters of tools and dies prefer it to other methods. Its use for production work is limited and yet it holds certain advantages which are difficult to obtain otherwise as may be realized when studying the data.

## Drawing

Sample No. 3 in Fig. 8 (0.60 per cent carbon) was chosen as a representative sample for studying the microstructures, and also for examining its behavior when drawn at various temperatures. This sample, time-quenched, had a surface hardness of approximately 51 Rockwell "C." By drawing at a suitable temperature it was

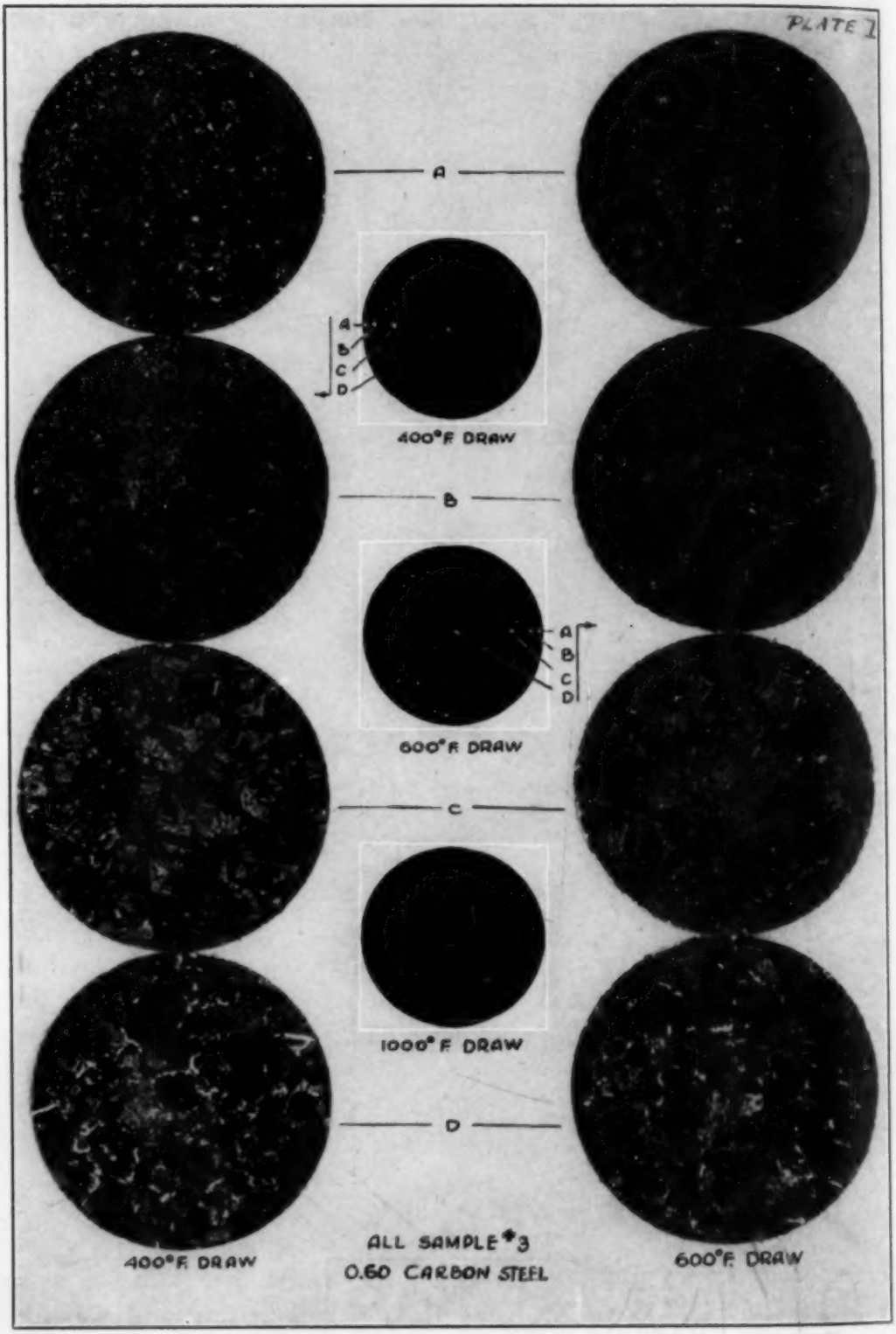


Fig. 12—Macrographs and Micrographs of 0.60 Per Cent Carbon Steel with Treatments as Shown.

possible to bring the higher subsurface hardness to a close approximation of the original surface hardness which did not change appreciably at these lower temperatures. This is shown in Fig. 11, which also indicates that in this case 600 degrees Fahr. produces a lowering of the subsurface peak to an approximate level of the surface hardness. It will also be noticed that the rest of the curve retains substantially the same form. After etching, the rings were still quite apparent after the various draws; however, a fading-out took place at 1000 degrees Fahr. (540 degrees Cent.).

A study of structures obtained after some of the above treatments is shown in Fig. 12. As would be expected when referring to the curves, the greater change takes place at the inner white ring (section "B"). Section "A" shows further tempering while the martensitic areas in all the other sections also start to break down at the lowest temperature used (400 degrees Fahr.). After heating to 600 degrees Fahr. sections "A" and "B" have a similar microscopic appearance, indicating that like hardness exists. This is shown by the curve to be true.

## Interrupted Quench

Heats 3A and 3B were used for this series of tests.

Water was used as the only quenching medium, being held to about 70 degrees Fahr. Time may be varied during any stage of the procedure, either during the first dip, during the interval in air, or during the last dip. We chose to vary the period in air, and knowing from previous information gathered by double quenching that five seconds in water produces a sharp dark outer ring, this period became a constant. The final dip lasted until the sample was completely cold.

The samples were prepared in the manner previously described, up to the time of quenching. Timing consisted of the following:

| Sample No. | Quenching Conditions |
|---|---|
| 1 | Water quench. |
| 2 | 5 seconds initial quench, 1 second in air, back into water. |
| 3 | 5 seconds initial quench, 3 seconds in air, back into water. |
| 4 | 5 seconds initial quench, 5 seconds in air, back into water. |
| 5 | Oil quench. |

The water and oil quenches were added in order to facilitate direct comparison of all samples.

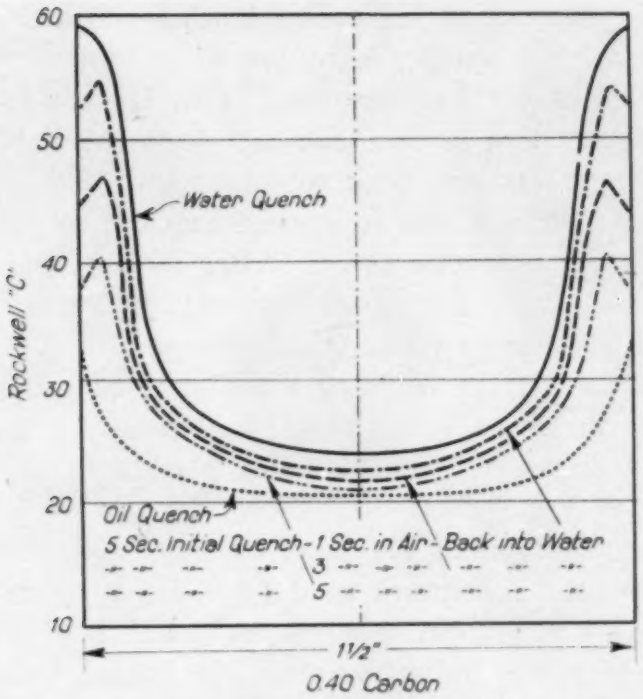


Fig. 13—Hardness Distribution Curves for Time-Quenched 0.40 Per Cent Carbon Steel.

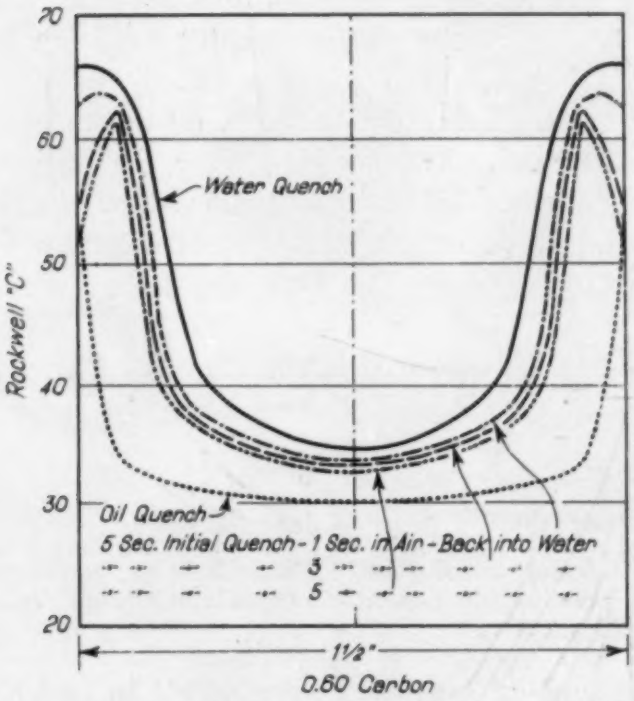


Fig. 14—Hardness Distribution Curves for Time-Quenched 0.60 Per Cent Carbon Steel.

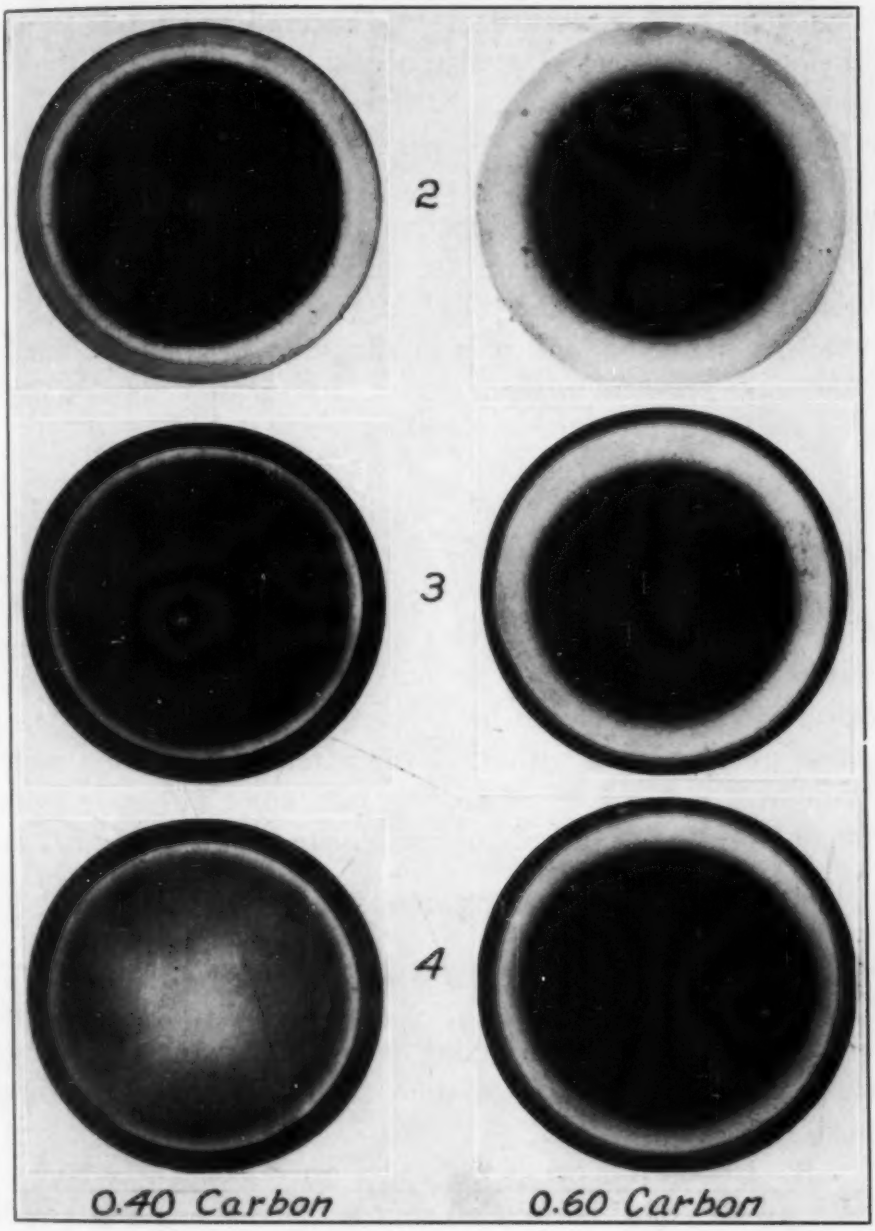


Fig. 15—Macrographs of Hardened and Etched Specimens of 0.40 and 0.60 Per Cent Carbon Steels. Figures Between Disks Refer to Sample Numbers.

The hardness distribution curves of the time-quenched samples shown in Figs. 13 and 14 indicate that subsurface hardness is consistently higher than surface hardness. This difference is even more notable here than on the previous curves produced by double quenching with a constant interval factor. The more drastic quench involved with this method as compared with the previous two methods is probably responsible for this fact. With increased interval of

time the actual surface hardness is lowered. This is expected; nevertheless the manner of occurrence is interesting. On the higher carbon steel the rate of surface hardness drop is faster than is true of the subsurface; on the 0.40 per cent carbon steel the rate of drop is about the same at these two positions. Apparently the difference in hardenability or the matter of critical combination of size and chemistry enters in again to cause this difference. Since a standard 1500 degrees Fahr. (815 degrees Cent.) quenching temperature was used for both carbon ranges, a slight advantage was given the 0.60 per cent carbon steel as its upper critical point was exceeded slightly more than on the 0.40 per cent carbon grade. This point appears of minor concern.

We see in Fig. 15 that the comparative thicknesses of the first two rings vary in the same order as noticed after double quenching. A certain similarity exists between results from double and interrupted quenching, although the greater similarity is in the appearance of the macro etch. An outstanding feature of varying the interval in air is the increase in drawing effect at the outer ring by conduction of heat from within. Even after a five-second interruption (sample No. 4) of a fast cooling rate, some austenite evidently still remains present to quench to martensite (the white inner ring).

## Conclusions

1. Three methods of time quenching have been analyzed qualitatively, namely:

A. Simple time quenching where the part is quenched in water for a given length of time, then removed and allowed to finish its cooling in air.

B. Double quenching where a part is quenched in water for a given length of time, then transferred immediately to oil to finish the quench.

C. The interrupted quench where the part is quenched for a given length of time in water, removed and exposed to the influence of air for a definite length of time, then returned to water to finish the quench.

2. It is to be noted that the conditions listed under "A", namely, the simple time quench, produce hardenability curves of the same general shape as straight oil or water quenching, whereas the conditions listed under "B" and "C", namely, double quenching and interrupted quenching, produce a differently shaped hardenability

curve from the straight oil or water quench in that under certain conditions the surface hardness is depressed with respect to the subsurface hardness.

3. The advantages of methods used in these series of tests lie principally:

A. In the comparatively tougher outer ring produced under the circumstances of lower stress that would be experienced with the straight water quench.

B. In obtaining of the necessary hardness with structural freedom from free ferrite in the outer part of the piece to be quenched, which would not be obtained with an oil quench.

4. Although time quenching should not be considered a panacea for problems of low hardenability oil quenching steel, or for parts showing a tendency to crack in water, its use can be widened by an understanding of the effect of the stepped cooling rate which results in the time quench.

5. Several practical examples of where the time quench has been used to advantage are illustrated.

6. The behavior of the steel under the conditions of time quenching has been explained on the basis of the "S" curve.

ACKNOWLEDGMENT

The authors wish to express their appreciation to Messrs. E. C. Smith and R. S. Archer for their comments and constructive criticism.

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## DISCUSSION

**Written Discussion:** By R. A. Grange, research laboratory, United States Steel Corp., Kearny, N. J.

Messrs. Burns and Brown have presented an interesting and timely paper dealing with the metallurgical aspects of quenching methods that have probably not heretofore received the attention their practical use would warrant.

Of particular interest to this writer was the application of the S-curve (Fig. 9) to explain the microstructures observed in double quenched sample No. 3 of the S.A.E. 1060 steel. An exact analysis on this basis, which is impossible without detailed experimental knowledge of the cooling history at the points "A", "B", "C" and "D", apparently was not intended by the authors. Nevertheless, in their schematic use of what is obviously the isothermal S-curve for eutectoid steel to explain the origin of "cooling transformation" structures

in a hypoeutectoid steel, the authors are neglecting factors which certainly deserve mention. It should be strongly emphasized that Fig. 9 is merely a rough, schematic chart drawn to facilitate the authors' version of the origin of the observed microstructures.

The structure at "A" (Fig. 8) is described by the authors as consisting of tempered martensite. Judging from the appearance of the photomicrograph taken at this point, the structure does consist principally of tempered martensite, but, in addition, contains about 10 per cent of a light-etching material that resembles untempered martensite. Apparently the water quench did not cause complete transformation of the austenite at "A" to martensite but left 10 per cent untransformed until after the final oil quench. The outer black ring of the macro-etched sample No. 3 therefore consists of a structure in which tempered martensite predominates but does not necessarily comprise the entire structure. Toward the center of the bar from "A", the structures decrease in the amount of tempered martensite present and increase in untempered martensite until the latter predominates and produces a light-etching macrostructure. Despite the authors' statement that the structure at "B" consists of martensite and troostite (fine pearlite), it appears more likely from the appearance of the micrograph taken at this point that the dark-etching constituent is tempered martensite, because of its acicular nature, rather than troostite, which would have formed a nodular network. Thus at "B" the structural components are the same as at "A" except that the amounts of each are approximately reversed. The structures in the inner portions of the white ring probably consist of martensite (untempered) and, as the border of the dark-etching core is approached, ever increasing amounts of fine pearlite (troostite) and ferrite. The white ring merges gradually into the dark core and appears white only so long as untempered martensite predominates.

In the major portion of the core of sample No. 3, Fig. 8, it is probable that transformation occurred only during the oil quench. The work of French* on the quenching of steel would indicate that, due to an inevitable cooling "lag" at the center of the bar, the austenite in the central area required considerably more than 5 seconds to cool to its transformation range. Therefore this austenite was probably not undercooled until the oil quench took effect and did not transform either during the water quench or during the time interval between quenches. Because of the effect of the initial water quench, the cooling rates through the significant Ar' region of the S-curve in this central area would undoubtedly be greater in the double quench than in a straight oil quench and conceivably might, in certain cases, exceed those resulting from a straight water quench. This latter point may explain the rather surprising fact that two of the hardness curves of Fig. 7 lie wholly, or in part, above the curve for the straight water-quenched bar.

**Written Discussion:** By A. F. Finkl, metallurgist, A. Finkl & Sons Co., Chicago.

The different structures and hardnesses produced when the cooling rate of steel is interrupted during the several types of quenching has been well demon-

*H. J. French, "A Study of the Quenching of Steels," TRANSACTIONS, American Society for Steel Treating, Vol. 17, 1930, p. 646-730.

strated by the authors. The photomacrographs accompanying these explanations are exceptionally fine.

After examining the photograph of the cracked die in Fig. 10, the question that presented itself was whether or not this die required a combination water and oil quench or had some few details been overlooked in the water hardening procedure.

The appearance of the cracks indicates the following:

(1) The strains from the forging process had not been removed by annealing before hardening.
(2) The die had been loaded into a furnace for hardening that was at too high a temperature.
(3) The rate of heating for quenching was too rapid.
(4) Not holding at quenching temperature long enough for the carbides to go into solution.

Any one or a combination of these conditions, in addition to the severe strains created by quenching, would be sufficient in a massive or irregularly shaped piece to cause relieve strain cracking.

The reason for discussing the cracks in this particular die is not to criticize the hardening of it, but rather to present an instance of a borderline case.

Oftentimes a massive or irregularly shaped piece of alloy steel will not harden properly in oil, but due to difficulty in handling or due to oil and water quenching baths being too far removed from one another, a combination water and oil quench would be impossible, therefore a straight water quench is necessary. If the above practices were considered and avoided many a hardening headache would be spared.

Considering the 0.60 per cent carbon steel sample mentioned in the double and also interrupted quench, it would be possible by improper timing of the quench or with a draw higher than 1000 degrees Fahr., used in Fig. 11, to create undesirable results under ordinary circumstances. The rate of softening of the martensitic surface is so much faster than that of the pearlitic core that the two curves, hardness vs. tempering time, at a given temperature actually cross. Thus, under appropriate quenching time and tempering ranges and time at temperature, the fully hardened martensite may be made softer than the pearlite which did not harden on quenching.

**Written Discussion:** By R. E. Cramer, special research assistant professor in engineering materials, University of Illinois, Urbana, Ill.

The writer is very much interested in Messrs. Burns and Brown's very practical paper on "Time Quenching". In my opinion some variation of their methods can be applied to many heat treating problems where quenching cracks are likely to develop. I recall a method developed in 1923 to harden several sizes of ball bearing races ranging from 6 to 10 inches in diameter. The races were heated in batch type furnaces and the quenching instructions were to dip the races in the circular quenching tank on a hook and circulate them around the circumference of the tank a certain number of times varying from about 6 to 10 times for the various sizes of races and then quickly remove the rings from the quenching water. Two or sometimes one more turn around the tank would practically always end in a cracked ring.

There is one part of Messrs. Burns and Brown's paper which applies to the writer's work on end-hardened railroad rails. That is the "5 seconds in water" curve of Fig. 2, page 212. It will be noted that there is a ring about $\frac{3}{16}$ inch from the surface which is about 3 or 4 Rockwell C numbers softer than the adjacent metal on the interior of the specimen. We have found this same thing under the hardened surface of many railroad rails which are end-hard-

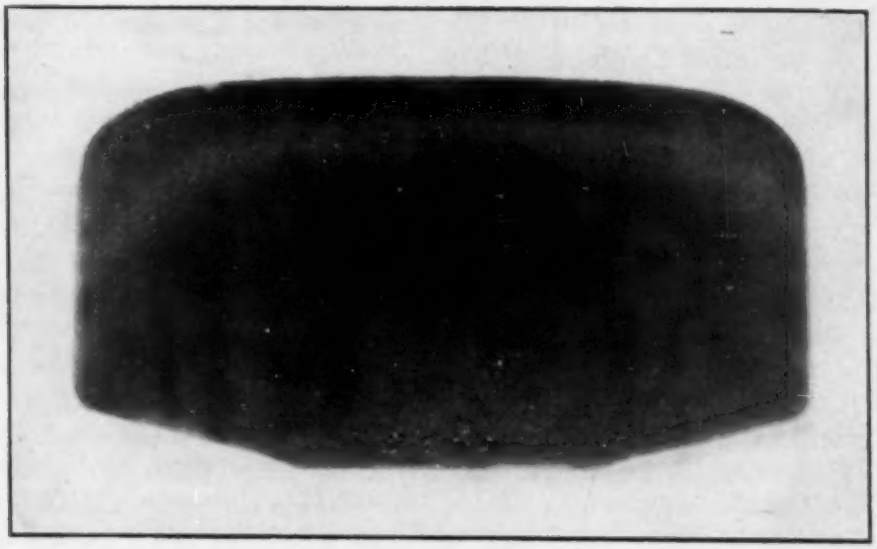

Fig. 1

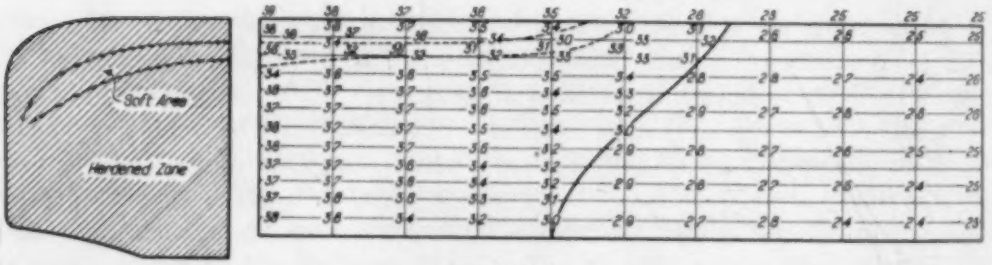


Fig. 2

ened from the rolling heat by a water spray directed on the surface of the rail head for about 30 seconds, followed by cooling in air. That is, there is a narrow zone underneath the hard surface which is softer than the interior metal below the soft zone. The accompanying Fig. 1 shows the soft zone as a light band parallel with the surface of the rail brought out by an ammonium persulphate etch. Fig. 2 shows the soft zone as indicated by Rockwell C readings on a vertical section through the head of an end-hardened rail.

I would like to ask the authors of this paper to explain the hardening mechanism which produces these soft zones in their round specimens and in the surface quenched rails.

**Written Discussion:** By E. F. Davis, metallurgist, Warner Gear Div., Borg-Warner Corp., Muncie, Ind.

Messrs. John Burns and Victor Brown have given us some concrete facts

on what occurs when time quenching is practiced and this constitutes a valuable addition to our knowledge of what transpires when this method of heat treating is used.

My comments will be more of a practical discussion, for I do not question the facts they have so ably presented.

The object of quenching is to arrest transformation, and the resulting phase depends upon the rapidity with which the heat is removed from the steel. A factor in this is the volume or mass, which enables it to store a greater or lesser number of heat units during the heating up period and which must be rapidly removed by the convection of quenching currents and rate of differential heat transfer between the steel and quenching medium. With a rapid removal of surface temperature these rings will be obtained. As the mass increases, the difficulty of obtaining similar macrographic and micrographic patterns would be increased and might show an entirely different result. In heat treating heavy sections, it is nearly impossible by any means of quenching to obtain the depth of hardness, physical properties, or impact values which may be obtained by heating the same steel in smaller sections. Another factor is inherent ability of the steel itself to harden due to the simplicity or complexity of its carbides, comparing, for example, an S.A.E. 1060 steel with a 0.60 per cent carbon high speed steel. The latter would harden to its center without even quenching.

The reason these points are mentioned is that any data which might be submitted would apply only to the specific type of steel under investigation, and moreover, only to the same diameter of specimen, and furthermore, only to the particular heat of steel investigated. A modification in the grain size, the presence of alloy residuals such as chromium, nickel and molybdenum, even in minute quantities, or a difference in the manganese content might produce an entirely different set of curves than those shown in Figs. 1, 2, 6, and 7. The carbon steels between 0.40 and 0.60 per cent carbon are particularly sensitive to these variables.

A skilled heat treater can do wonders with the double quench, and this was formerly a common practice in tool hardening, and, as the authors intimate, is still practiced in the industry. The avoidance of double quenching is in my mind more desirable than its performance. Steels should be selected which would eliminate the necessity for it. A double quench is bad due to the contamination of water in the oil bath—and I speak from experience rather than opinion. Water in quenching oil is responsible for much of the distortion and cracking in tool and production hardening. Another disadvantage is the human element. The simpler the quenching system, the better the results. In these days of mechanized heat treatment, I question its value in production heat treatments.

**Written Discussion:** By A. S. Jameson, metallurgist, International Harvester Co., Chicago.

We have little to say of the data and method of acquiring it. The test results covering quenching methods which have been and are to some extent still practiced, are well arranged and presented.

Time quenching is a makeshift method of increasing the physical properties

of rolled steel. In the absence of satisfactory intermediatory quenching mediums, parts should be designed to take a water or an oil quench, or the problem approached by assigning the proper steel analysis to a part where the design has been dictated by mechanical considerations only. Certainly, the maximum physical properties for a given hardness are not obtained from one of the steels used by the authors, S.A.E. 1040, by time quenching. It is bad enough to have to contend with mass conditions which govern cooling rates without introducing a quenching method which adds further complications.

The principles of the heat treatment of S.A.E. 1040 steel for the maximum physical properties at a given hardness consist, first, of obtaining a uniform austenite and second, of cooling rapidly enough to prevent agglomeration of the excess constituent, ferrite. The physical properties of the quenched and tempered steel depend on the degree to which these principles have been followed.

The following data will illustrate how the physical properties can vary with the same hardness, in a number of locations in a time quenched specimen at a hardness of 228 Brinell. Let us suppose one section cooled to 228 Brinell directly. The structure would appear microscopically to consist of about equal parts of ferrite and sorbite with tensile properties as follows:

| Yield Point | Tensile Strength | Elongation | Reduction of Area | Ratio YP to TS |
|---|---|---|---|---|
| 80,000 | 115,000 | 27 | 63 | 69 |

Another section cooled to 255 Brinell and was tempered to 228 Brinell. The microstructure would consist of about 30 per cent ferrite and 60 per cent sorbite. The tensile properties would be—

| Yield Point | Tensile Strength | Elongation | Reduction of Area | Ratio YP to TS |
|---|---|---|---|---|
| 91,000 | 122,000 | 25 | 62 | 74 |

A third section cooled to 600 Brinell and was tempered to 228 Brinell. The structure would then consist of sorbite with these properties:

| Yield Point | Tensile Strength | Elongation | Reduction of Area | Ratio YP to TS |
|---|---|---|---|---|
| 107,000 | 119,000 | 24 | 65 | 90 |

The presence of various amounts of structurally free ferrite produced by the various cooling rates effects the yield point but none of the other properties as recorded by the tensile test.

This might be unimportant but what might be important would be the detrimental effect of the free ferrite on the machining properties should machining be necessary on the finished part.

This and recent studies illustrate the complexity of the problem of heat treating hypoeutectoid steels. Our previous conception should no longer exist, that a tension, torsion or impact test as obtained mainly on specimens of these steels whose properties varied throughout the section tested, were more than a poor average estimate. The attainment of structural homogeneity in a section of heat treated hypoeutectoid steel is a problem that can be solved only by a study of a subject which is referred to as "Hardenability". Physical test data on these steels can only be acquired when homogeneity is attained.

The foregoing is not to be construed as a criticism of the paper as the authors hold no brief for time quenching which it must be admitted in many cases serves a useful purpose.

**Written Discussion:** By W. H. Mayo, Columbia University, New York City.

Published investigations of the type presented in this paper are of value because they emphasize our incomplete knowledge of the austenite decomposition in even the common carbon steels.

The photomicrograph of the outer ring, "A", shown in Fig. 8 appears to have, in addition to a matrix of tempered martensite, numerous small areas of white, untempered martensite. Selective tempering of martensite in the field shown seems very improbable. Rather, one would believe that this outer ring had, on the initial quench, only the opportunity to partly transform to martensite before being drawn back during the interruption in the quench, this heat tempering that martensite; then, when the sample was quenched in oil, the remaining austenite transformed to the white martensite shown in the photomicrograph.

Attention should be called to the presence of a few dark needles of bainite (pseudo-martensite) in the photomicrograph of ring "B" in Fig. 8. This serves to emphasize the authors' point that "the transformation is progressive and not instantaneous."

The authors call our attention to the fact that the dark outer ring "A" is thicker on the 0.40 per cent carbon steel than on the 0.60 per cent carbon steel in both Figs. 8 and 15, but they have little to say concerning this point. The phenomenon is apparently independent of the hardenabilities of the steels, for the straight water quench shows that the 0.60 per cent carbon steel has a higher hardenability than the 0.40 per cent carbon steel. The "A" ring is principally tempered martensite and it represents the depth to which martensite began to form before the quench was interrupted. If we consider any two samples in Fig. 8 that have the same thermal history then corresponding points in these samples were at the same temperature at any given time. From this it follows directly that the martensite in the "A" ring must have started to form at a higher temperature in the 0.40 per cent carbon steel than in the 0.60 per cent carbon steel or that the martensite transformation itself proceeded more rapidly in the lower carbon steel once it began. Both of these facts are known to us and have been demonstrated by Davenport and Bain and by Carpenter and Robertson.

One cannot help but wish that the authors had repeated the time cycle of their water-oil quench when they performed the interrupted water quench and so have shed more light on the anomaly of the two interrupted water-oil quench curves which lie inside of the straight water quench curve in Figs. 6 and 7.

### Authors' Reply

Mr. Cramer's experience with time quenching of ball bearing races and end quenching of railroad rails is very interesting. His comments on a softer subsurface area at the quenched end of the rail does appear to have a decided similarity to the softer ring on sample No. 2 of Fig. 2 which received a 5-second quench in water. Both the photomacrograph (Fig. 1 of the discussion)

and the plot of hardness shown in Fig. 2 of the discussion reveal the presence of the soft area. A definite explanation cannot be made for the procedure of transformation which causes this condition without further work. Interesting results may be produced if it is possible to plot a time-temperature curve at such a section during cooling by timed quenching.

In reply to Mr. Finkl's discussion, to the authors' knowledge the die shown in Fig. 10 was thoroughly annealed before hardening, then placed in a cold furnace and heated slowly and completely before quenching. Referring to point (4) of the discussion it is not necessary to obtain complete solution of carbides provided a uniform temperature throughout the mass is achieved so that the carbides out of solution are uniformly distributed. It is the aim sometimes to retain carbides out of solution where it is found desirable to curtail the hardening response of a steel where the probability of cracking is high.

It is possible to visualize a condition where a surface composed of drawn martensite could be softer than a center composed of primary cooled structures but in such an instance the general hardness would be quite low and application of such a part would be made accordingly.

Mr. Davis' comments concerning mass increase, quenching and their relation to physical properties obtainable are correct. As mass is increased it will be necessary to make adjustments as to hardenability of the steel. Any slight change caused by a variation in hardenability, whether due to chemistry or grain size, and/or size of the bar will produce a difference in the curve. Concerning the bar size, the smaller the mass involved the more sensitive the part will be to slight variations in time; this introduces a size limit to which time quenching may be applied effectively. Mr. Davis' reference to the sensitivity of steels between 0.40 and 0.60 per cent carbon is well appreciated and this was a factor in their choice for the work in this paper.

Mr. Davis' comments on the application of the double quench to tool hardening deserve consideration, especially his experience with water in the oil tank. It is obvious that such an occurrence might well prove disastrous, especially when parts inherently sensitive to quenching should be involved.

Mr. Grange is correct in stressing the fact that Fig. 9 is a rough sketch of an S-curve and that it is employed merely as a means to illustrate the points brought out in the discussion of microstructures as related to the cooling of the samples.

Mr. Grange's observation as to the structure in ring "A" is in agreement with the authors' in that it is largely composed of tempered martensite. However, it is felt that the presence of a small amount of untempered martensite may or may not indicate a slight amount of incomplete transformation. The period in air of approximately one second at the most is a very short one for drawing and which may therefore result in only a partially drawn structure. The authors feel that incomplete transformation is the most probable explanation but the other possibility must not be overlooked.

Mr. Grange's observations regarding the tapering of the various structures are well taken. We wish to point out that a difference of nomenclature has led to some misunderstanding since Mr. Grange's term "tempered martensite" and the authors' term "troostite" denote the same thing. This difference will be

cleared up. The reproduction in the preprint is not clear but some of the dark etching material just within the light ring "B" is an acicular structure of pseudo-martensite.

It is quite logical to assume that a substantial amount of transformation through the core section took place during the oil quench, sample No. 3, Fig. 8, but the presence of a considerable amount of pseudo-martensite through "C" raises the question of possible constant temperature transformation, i.e., whether under certain conditions of timing it may be possible to promote such transformation—at least to a limited extent. If this be true then it appears that it would occur during the interval in air. This section is not far removed from the surface. On the other hand, the central portion obviously cools at a far slower rate and is affected relatively less by the water dip than those sections nearer the surface.

Mr. Mayo's discussion concerning the matter of selective tempering is similar to a previous one which is covered above.

Referring to Figs. 8 and 15 and to the fact that the dark outer ring "A" is thicker on the 0.40 per cent carbon steel than on the 0.60 per cent carbon steel, hardenability and depressed thermal transformation are related in these carbon steels and are controlled by the chemistry. Therefore this dark outer ring may be varied in thickness by either varying the chemistry or the timing or both.

Mr. Jameson's view concerning time quenching in general versus the simpler straight oil or water quenching is a practical one and is shared by all of us for the great bulk of production parts. However there are times when it becomes necessary to resort to "makeshift" methods and at such times we welcome information relating to our problem.

The authors wish to express their appreciation to the writers of the discussions for their interest and valuable contributions to the work on time quenching. The subject is broad and many questions as well as possibilities have presented themselves. Further effort should prove fruitful not only in clearing up the practical problems and questions but also in throwing light on various technical points.

# THE FATIGUE RESISTANCE OF STEEL AS AFFECTED BY ACID PICKLING

By G. L. Kehl and C. M. Offenhauer

*Abstract*

*This paper presents results of an investigation concerning the fatigue properties of a 0.54 per cent plain carbon steel as affected by prior stressless corrosion in a sulphuric acid pickling bath. Data is presented herewith showing the influence on fatigue properties brought about by varying, within commercial limits, a selected "standard" pickling procedure. The variables so investigated were the effect of certain inhibitors, time of pickling in both inhibited and uninhibited baths; temperature, acid concentration, and ferrous sulphate content of the bath.*

*Complete fatigue curves are presented and in some cases plotted relationships are shown between cycles to failure and other variables at one particular stress. An attempt has been made to correlate fatigue properties with the surface notch condition of the steel that resulted from the various controlled deviations of the pickling procedure. Photomicrographs of the average surface condition of the metal for all series of tests are included.*

DURING the past four years investigations have been carried out at Lehigh University on the surface treatment of metals. The purpose has been one of securing good surface cleaning and satisfactory plating conditions with the least detrimental effect on the endurance strength of the cleaned or cleaned and plated material.

In a previous report by one of the authors (1)[1] the effects on the endurance life of steel resulting from such mechanical cleaning methods as sand, steel grit, and steel shot blasting, have been reported. Two general, fundamental effects were shown in this investigation. Blasting produces surface notches and pits which are

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

The material in this paper has been abstracted from a thesis submitted by C. M. Offenhauer in partial fulfillment of the requirements for the degree of Master of Science, Lehigh University, 1939.

A paper presented before the Twenty-first Annual Convention of the Society held in Chicago, October 23 to 27, 1939. Of the authors, G. L. Kehl was formerly instructor in metallurgy, Lehigh University, Bethlehem, Pa., and C. M. Offenhauer was formerly New Jersey Zinc Company Research Fellow, Lehigh University, Bethlehem, Pa. Manuscript received June 23, 1939.

decidedly deleterious to good fatigue properties whereas the surface cold working resulting from impingement of the blasting medium has the effect of improving fatigue properties. By the proper choice of blasting medium, blast pressure, and angle of blasting, it was shown that the fatigue properties could be made to equal, and exceed, the fatigue properties of the material before cleaning.

Acid pickling, generally in dilute sulphuric acid solutions, is one of the most common methods employed for cleaning steel surfaces. It is conceived that acid pickling, unlike mechanical cleaning methods, will produce only deleterious effects with regard to fatigue properties, namely those due to the pitted surface resulting from the acid attack and the possible deleterious effects caused by hydrogen embrittlement. The beneficial effects due to a work-hardened surface, as characteristically produced by the mechanical cleaning methods mentioned heretofore, are absent in acid pickling. It would appear, therefore, that under no conditions of pickling would it be possible to raise the fatigue limit above that of the material "as-received"; but by choice of proper pickling temperatures, acid concentration of the bath, and type of inhibitors that might be used, this limit may be approached by the pickled material.

This paper is a report on the fatigue properties of steel as affected by acid pickling methods, wherein the resulting fatigue properties are correlated with the acid concentration of the bath, time and temperature of pickling, and other important factors of the procedure.

## Apparatus and Experimental Procedure

*Fatigue Machine*—The fatigue machine used in this investigation, shown in Fig. 1, was of the reverse bending type. Detailed information with regard to actual operation of the machine may be obtained elsewhere (1).

*Specimens*—The specimens employed in this investigation were contributed by John A. Roebling's Sons Company through the courtesy of Wilber E. Harvey. The specimens were obtained by cold rolling a heat treated rod to cross sectional dimensions of ⅛ inch x ½ inch. The flat wire was then continuously heat treated at 1600 degrees Fahr. (870 degrees Cent.) and quenched in molten lead at 900 degrees Fahr. (480 degrees Cent.). This final heat treatment produced a fine pearlitic (sorbitic) structure. After the 900 degrees Fahr. lead quench, the wire was passed through straight-

ening rolls to effect any straightening that might be necessary, and finally cut into convenient 24-inch lengths. In the investigation actual lengths of the specimens varied between 12.38 to 13.7 inches.

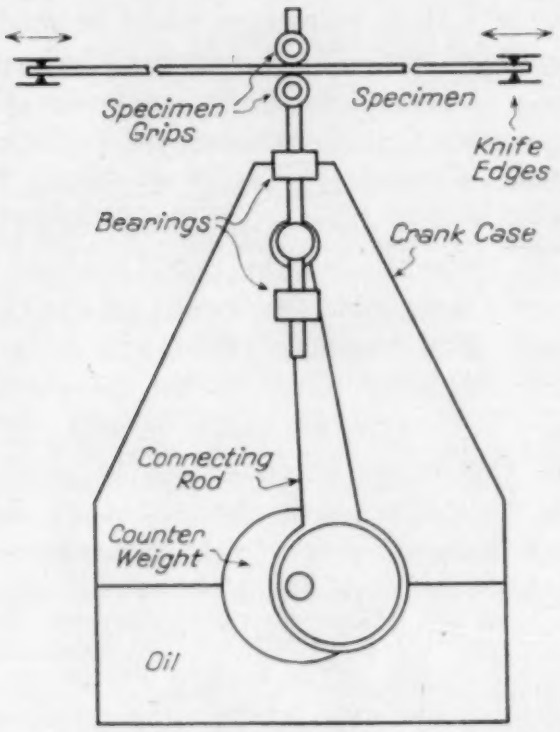


Fig. 1—Cross-sectional View of Fatigue Machine.

It was reasonable to believe that the specimens were not appreciably work-hardened during straightening to the extent that such work-hardening would result in nonuniform fatigue properties. Frye and Kehl (1), using identical specimens, concluded by Vickers hardness measurements that the straightening rolls produced no measurably hardened layer.

To determine whether or not the heat treating operation produced any decarburization, the presence of which would be detrimental to fatigue life, a number of specimens in the "as-received" condition were examined microscopically. No decarburization was discernible at 1000 diameters.

The chemical composition and physical properties of the material "as-received" are shown below.

| Chemical Composition—Per Cent | | | | |
|---|---|---|---|---|
| C | Mn | P | S | Si |
| 0.54 | 0.49 | 0.024 | 0.017 | 0.182 |

| Physical Properties | |
|---|---|
| Tensile strength ..... | 136,000 pounds per square inch |
| Yield strength ..... | 91,000 pounds per square inch |
| Elongation in 2 inches..... | 13.5 per cent |
| Reduction in area ..... | 37.0 per cent |
| Modulus of elasticity ..... | 29,400,000 pounds per square inch |

*Pickling Apparatus*—The pickling equipment used in this investigation consisted of a large earthenware tank supported within a sheet iron water bath. The water bath was heated by a set of gas burners, and by manual manipulation of these burners it was possible to maintain the temperature of the pickling medium within ±2 degrees Fahr. During pickling moderate circulation of the bath was secured by means of a laboratory stirring motor (Fig. 1A).

The specimens undergoing pickling were supported, out of contact with each other, on two U-shaped glass rods suspended from the edges of the tank. This arrangement facilitated the removal of all specimens simultaneously from the bath and permitted free circulation of the pickling medium around each specimen.

*Pickling Procedure*—All pickling solutions employed in this work were prepared from deaërated distilled water and 66-degree Baumé sulphuric acid (c.p.) and other additions as subsequently noted. A "standard" pickling procedure with regard to acid concentration, time of pickling, and temperature of the solution was arbitrarily established after considering the range of these variables as recommended in the literature (2). The "standard" pickling bath finally adopted consisted of an aqueous solution of sulphuric acid (10 per cent by weight), used at a temperature of 149 degrees Fahr. (65 degrees Cent.). The "standard" time of pickling was 12 minutes. In the subsequent work, each of these selected "standards" was separately varied within commercial limits.

At the conclusion of each pickling operation, the cleaned specimens were removed simultaneously from the bath and immediately immersed in a large volume of water. This water rinse was followed by a final rinsing, and swabbing with soft cotton, in ethyl alcohol to remove the last traces of scum and smut. Drying of the specimens was expedited by a blast of warm air.

This procedure of cleansing the specimens after the acid pickle rather than by the conventional lime wash was adopted in order to eliminate any surface variables that may have resulted from reaction between the lime and adhering acid. It was determined early in this investigation that the procedure followed was amply effective in preventing further surface corrosion.

To insure practically complete removal of hydrogen embrittlement and accompanying temporary deleterious effects, all specimens were allowed to stand at room temperature for 72 hours before being tested. This procedure is in accord with the findings of Langdon and Grossmann (3).

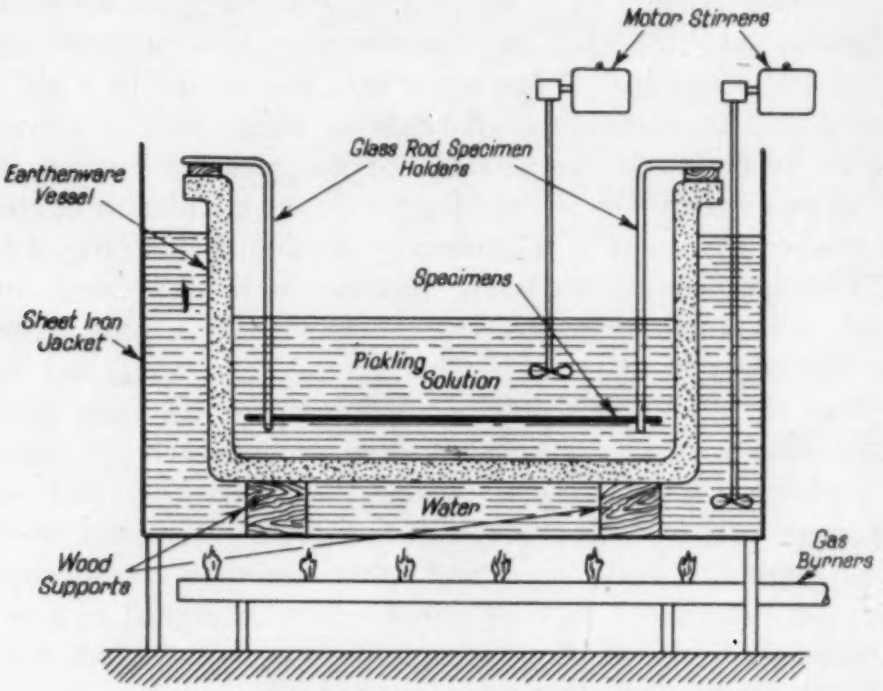


Fig. 1A—Sketch of Pickling Test Apparatus.

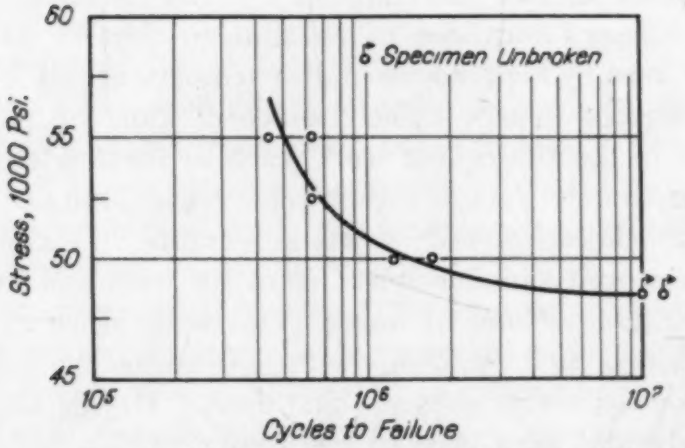


Fig. 2—S-N Curve of Material "As-Received."

## Resulting Fatigue Properties

*Material "As-Received"*—To establish a basis upon which to compare the effects of the different pickling procedures on the fatigue properties of the material, the fatigue limit of the material

Fig. 3—Surface Notch Condition of Material "As-Received." × 260.

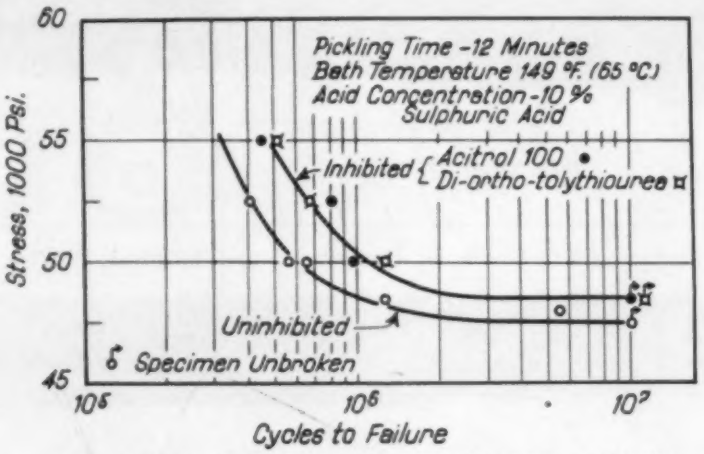


Fig. 4—S-N Curve of Material Pickled With and Without Inhibitors.

"as-received" was determined, as shown in Fig. 2. The fatigue limit was found to be 48,500 pounds per square inch or approximately 35 per cent of the tensile strength.

In an attempt to correlate fatigue properties with the size and shape of notches resulting from pickling, microscopic studies were made of each group of specimens treated. The surface condition of the material "as-received" is shown in Fig. 3. This photomicrograph and all subsequent ones, as will be noted later, were prepared from an edge of the specimen, parallel to the longitudinal axis. At least four inches of a specimen from each group of treated specimens were microscopically examined at 350 diameters (reduced to 260 diameters in reproducing). In the opinion of the authors, the photomicrographs presented herewith represent, in general, a qualitative indication of the actual notch condition.

*"Standard" Acid Pickle*—The fatigue limit resulting from pickling for 12 minutes in an aqueous solution of sulphuric acid (10 per cent by weight) at a temperature of 149 degrees Fahr. (65 degrees Cent.) is shown in Fig. 4. It will be noted, in comparison to Fig. 2, that the acid pickling caused a decrease in fatigue limit of

Fig. 5—Surface Notch Condition Resulting From "Standard" Acid Pickling. × 260.

Fig. 6—Surface Notch Condition Resulting From Inhibited Pickling in Di-ortho-tolythiourea. × 260.

Fig. 7—Surface Notch Condition Resulting From Inhibited Pickling in Acitrol-100. × 260.

1000 pounds per square inch, or approximately 2 per cent. This decrease is due, in all probability, to the somewhat more severe notch condition resulting from the acid pickle, as shown in Fig. 5.

*"Standard" Acid Pickle with Addition Agents*—To study the effects on the fatigue properties of the protection afforded the surface of a metal by the addition of inhibitors to the pickling bath, two groups of specimens were pickled in different solutions, one containing approximately 0.033 per cent by weight of Di-ortho-tolythiourea,[2] and the other 0.023 per cent by weight of Acitrol-100.[3] The pickling time and temperature of the bath was maintained according to the "standard" procedure.

It will be noted in Fig. 4 that both inhibitors are equally effective in preventing a decrease in fatigue properties, as both curves show a fatigue limit (48,500 pounds per square inch) equal to that of the material "as-received." Although the specimens were cleaned

[2]A recommended inhibitor. W. Machu, *Transactions*, American Electrochemical Society. Vol. 47, 1936.

[3]A commercial inhibitor. Amount used recommended by manufacturer.

equally as well as those pickled in the "standard" uninhibited pickling bath, the apparent absence of damage resulting from the inhibited pickling may be attributed to the resulting less severe notch condition. This, we believe, is confirmed in Figs. 6 and 7.

*Effect of Pickling Time—"Standard" Uninhibited Pickling Bath*—It would be reasonable to believe that the longer the time of pickling, other factors remaining constant, the greater will be the

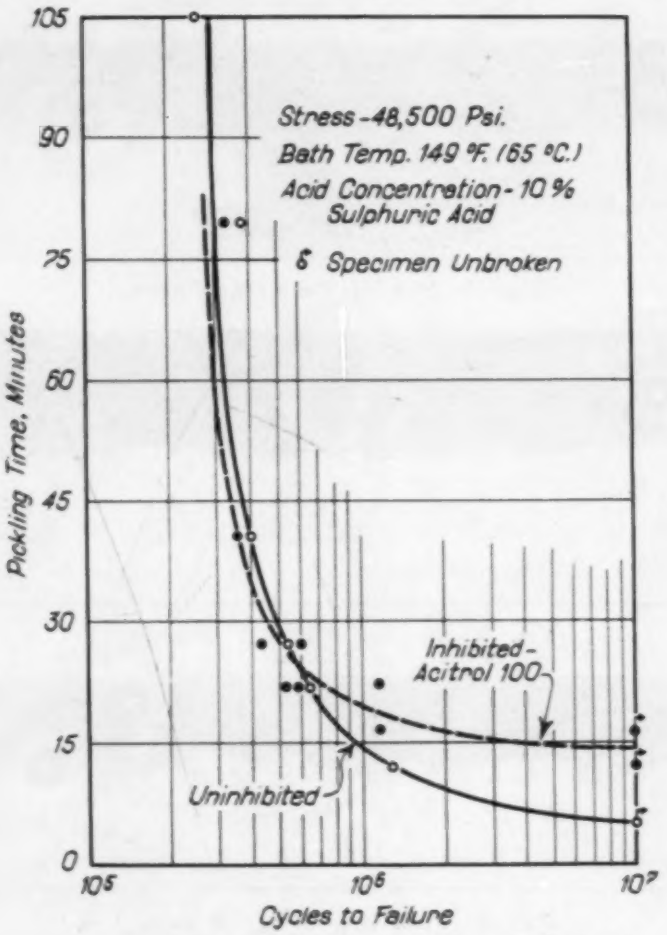


Fig. 8—Effect of Pickling Time on Fatigue Life.

surface corrosion of the metal with a corresponding decrease in fatigue life.

To study the effects of pickling times on the fatigue properties of the metal, six series of specimens were pickled in the "standard" bath for times ranging between 5 and 105 minutes.

The relation between fatigue life at a stress of 48,500 pounds per square inch and time of pickling is shown in Fig. 8.

It is of interest to note in Fig. 8 that for immersion times be-

Fig. 9a—Pickling Time—5 Minutes.

Fig. 9b—Pickling Time—12 Minutes.

Fig. 9c—Pickling Time—17 Minutes.

Fig. 9d—Pickling Time—22 Minutes.

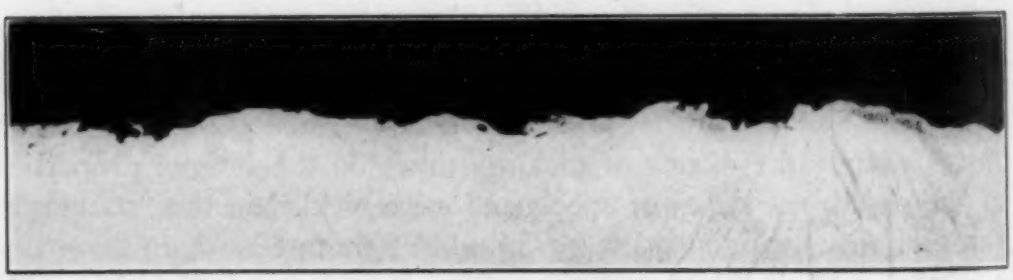

Fig. 9e—Pickling Time—27 Minutes.

Fig. 9—Surface Notch Condition Resulting From Uninhibited Pickling for Various Lengths of Time. × 260.

Fig. 9f—Pickling Time—40 Minutes.

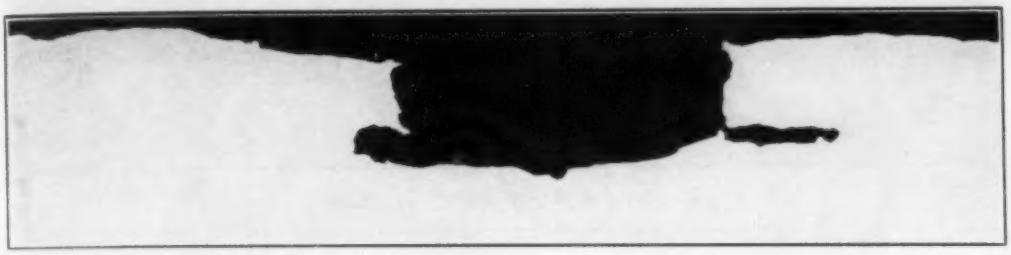

Fig. 9g—Pickling Time—80 Minutes.

Fig. 9—Surface Notch Condition Resulting From Uninhibited Pickling for Various Lengths of Time. × 260.

tween 5 and 30 minutes the fatigue life of the metal decreases relatively rapidly as the time of pickling is increased; whereas for immersion times greater than 30 minutes (in this case between 30 and 105 minutes) increased corrosion of specimens has apparently relatively little effect on the fatigue properties.

The photomicrographs shown in Figs. 9a to Fig. 9g indicate a general trend from a number of small surface pits formed at relatively short times of immersion to a few relatively large, well rounded pits formed at comparatively long times of immersion. Presentation of the intensity of notches resulting from the 105-minute pickle has been omitted because the resulting surface corrosion was found to be so severe that it was impossible to select a representative notch condition.

The results of this part of the investigation are more or less similar to the findings of Kandler and Schulz (4). They found that specimens which were mechanically "V" notched, followed by acid corrosion in hydrochloric and nitric acids for various lengths of time, actually showed an increase in repeated impact fatigue life, up to a certain limit, as the time of corrosion increased. This increase was shown to be primarily due to the rounding off, during

Fig. 10a—Pickling Time—17 Minutes.

Fig. 10b—Pickling Time—22 Minutes.

Fig. 10c—Pickling Time—27 Minutes.

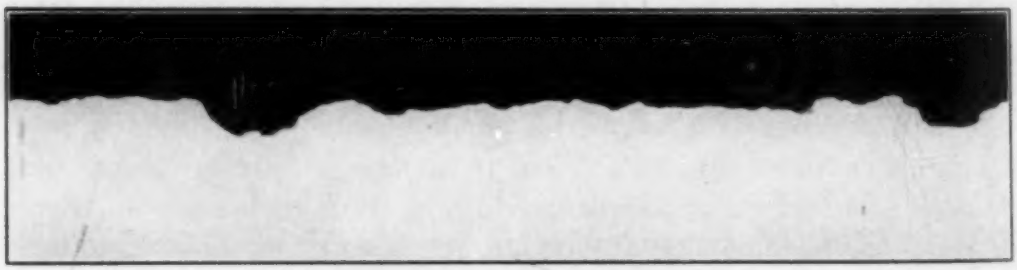

Fig. 10d—Pickling Time—40 Minutes.

Fig. 10e—Pickling Time—80 Minutes.

Fig. 10—Surface Notch Condition Resulting From Inhibited Pickling in Acitrol-100 for Various Lengths of Time. × 260.

corrosion, of the sharp mechanically formed notches, the rounding effect becoming decidedly more pronounced as the time of corrosion was prolonged.

*Effect of Pickling Time—"Standard" Inhibited Pickling Bath*—It will be noted in Fig. 8 that the addition of Acitrol-100* to the "standard" pickling bath showed a relationship between fatigue properties and time of immersion that was similar to that obtained from pickling in the "standard" uninhibited bath. A deviation of the two curves is noted, however, at the shorter times of immersion up to about 22 minutes, where the inhibited pickled specimens showed

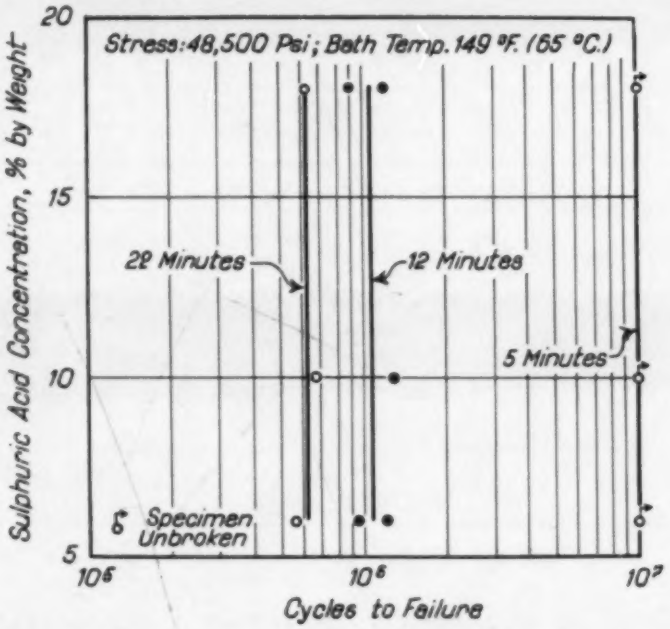


Fig. 11—Effect of Sulphuric Acid Concentration on Fatigue Life.

somewhat better fatigue properties than those pickled in an uninhibited bath. For times greater than 22 minutes, the curves practically coincided.

The notch condition of the surface resulting from inhibited pickling is shown in Figs. 10a to 10e. It will be noted that the trend of severity and intensity of notches follows very closely those resulting from uninhibited pickling (Figs. 9a to 9g).

*Effect of Acid Concentration*—In commercial pickling practice, the acid concentration of the bath is varied over wide limits (approximately 2.5 to 20 per cent depending upon type of scale and final surface desired. To determine what effect acid concentration might have on the fatigue properties of acid pickled material, specimens were

*0.023 per cent by weight, as recommended by the manufacturer.

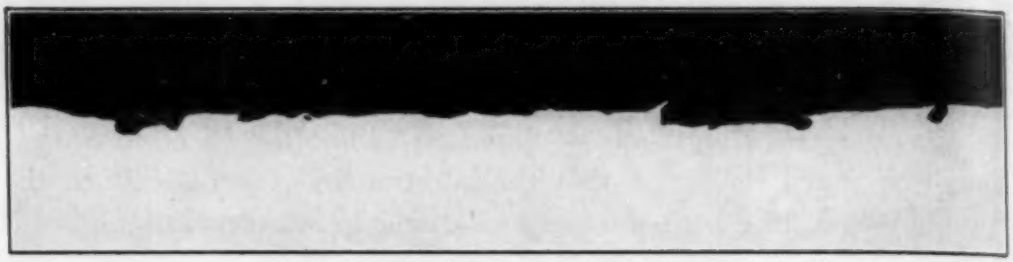

Fig. 12a—6 Per Cent Sulphuric Acid.

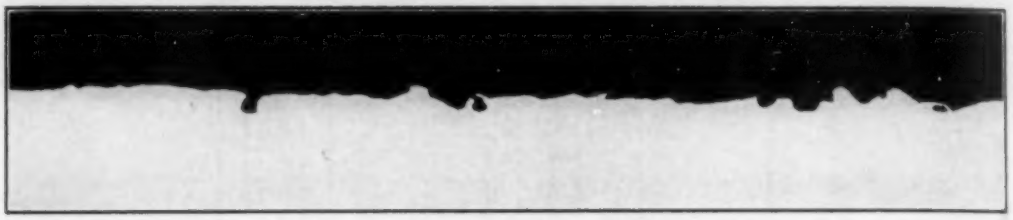

Fig. 12b—10 Per Cent Sulphuric Acid.

Fig. 12c—18 Per Cent Sulphuric Acid.

Fig. 12—Surface Notch Condition Resulting From Uninhibited Pickling for 12 Minutes in Baths of Different Acid Concentration. × 260.

pickled for three different lengths of time (5, 12, and 22 minutes) in three different pickling baths containing by weight 6, 10, and 18 per cent sulphuric acid. Pickling was carried out at a temperature of 149 degrees Fahr. (65 degrees Cent.). The results of this work are shown in Fig. 11.

It is of interest to note that for any one particular time of immersion the fatigue properties are independent of the acid concentration in the range investigated. As would normally be expected and as shown in Fig. 11, the shorter the time of immersion, the better the resulting fatigue properties.

The notch condition resulting from pickling for 12 minutes in the different acid solutions is shown in Figs. 12a to 12c. Photomicro-

graphs of the surface condition of the 5- and 22-minute pickled specimens have been omitted as the general character of the notches is similar to those shown in Fig. 12.

*Effect of Bath Temperature*—In commercial pickling practice the temperature of the pickling bath, as the acid concentration, may be varied over wide limits to meet certain pickling requirements.

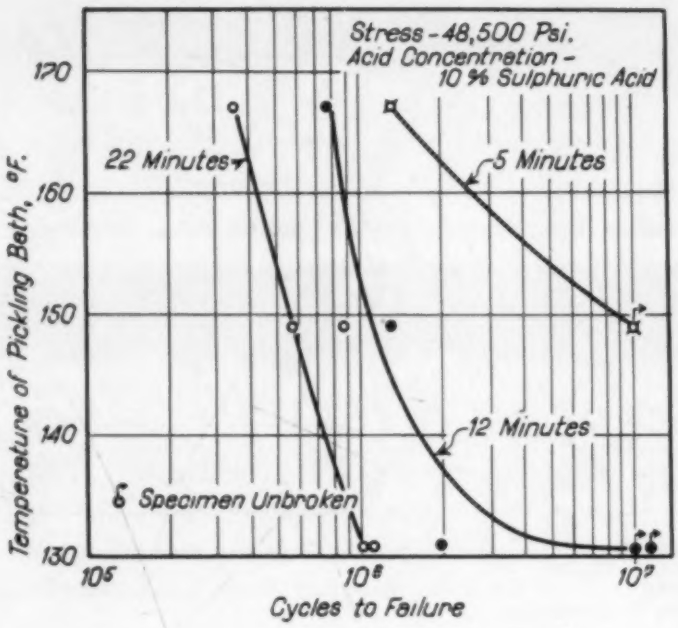


Fig. 13—Effect of Pickling Bath Temperature on Fatigue Life. Uninhibited Bath.

These limits may vary from approximately 130 to 212 degrees Fahr. (55 to 100 degrees Cent.). From consideration of reaction rates, it would be expected that an increase in the temperature of the bath (time of pickling constant) would increase general surface corrosion of the metal and, in turn, decrease fatigue properties.

To investigate this problem, specimens were pickled in the "standard" bath at temperatures of 131, 149, and 167 degrees Fahr. (55, 65, and 75 degrees Cent.). For each temperature investigated, specimens were pickled for 5, 12, and 22 minutes.

It will be noted in Fig. 13 that for each particular time of immersion the fatigue properties are markedly decreased with increase in temperature of the bath. This drop in fatigue life is in all probability due to the greater severity of surface notches resulting from increase in bath temperature, as shown in Figs. 14a to Fig. 14c.

Fig. 14a—Pickling Bath Temperature 131 Degrees Fahr. (55 Degrees Cent.).

Fig. 14b—Pickling Bath Temperature 149 Degrees Fahr. (65 Degrees Cent.).

Fig. 14c—Pickling Bath Temperature 167 Degrees Fahr. (75 Degrees Cent.).

Fig. 14—Surface Notch Condition Resulting From Uninhibited Pickling for 12 Minutes at Various Temperatures. × 260.

*Effect of Ferrous Sulphate Content of the Bath*—To determine what effect the ferrous sulphate content of the pickling bath might have on the fatigue properties, specimens were pickled for 12 minutes in baths containing, by weight, 5, 10, and 15 per cent ferrous sulphate. The fatigue life obtained at stresses of 48,500 and 50,000 pounds per square inch respectively were so inconsistent and the data so scattered that the results are not presented here. As far as this part of the investigation was conducted, there appeared to be no correlation between ferrous sulphate content and fatigue life.

## Conclusions

Inasmuch as variations of acid concentration of the pickling bath, selection of inhibitors, pickling times, and temperature of the bath are unlimited, and since the type and thickness of mill scale adhering to the steel influences the selection of the aforesaid, no universally valid conclusions can be drawn from this investigation. How-

ever, even though the data are incomplete in this respect, the pickling procedures within commercial limits and variations therein, for this particular steel, strongly favor the following views:

1. Pickling under "standard" conditions, i.e., 12 minutes in a 10 per cent (by weight) aqueous solution of sulphuric acid at a temperature of 149 degrees Fahr. (65 degrees Cent.) reduces the fatigue limit of the material "as-received" by about 2 per cent.

2. The addition of certain inhibitors to the pickling bath, namely Di-ortho-tolythiourea and Acitrol-100, minimizes the formation of deleterious surface notches and prevents a decrease in fatigue properties from those of the material "as-received."

3. The fatigue properties of specimens decrease as the time of pickling increases in uninhibited baths. This decrease in fatigue properties is most pronounced from immersion times of 5 to 22 minutes. Longer times of immersion up to a maximum of 105 minutes result in nearly constant fatigue properties throughout this range.

4. Throughout the range of pickling times of 5 to 15 minutes, specimens pickled in an Acitrol-100 inhibited bath show nearly constant fatigue properties. For immersion times greater than 22 minutes (up to 80 minutes), the fatigue properties are practically the same as specimens pickled for the same lengths of time in an uninhibited bath.

5. The fatigue life of the material is, for all practical purposes, independent of the acid concentration of the pickling bath within the range of concentrations investigated (5 to 20 per cent), other conditions of time and temperature remaining constant.

6. Increase in temperature of the pickling bath, between 131 and 167 degrees Fahr. (55 and 75 degrees Cent.), result in a pronounced decrease in fatigue properties.

It may be said, in general, that the results of this investigation indicate that the effect of prior stressless corrosion in acid in relation to resulting fatigue properties is primarily an effect of chemically formed notches. The effects of other factors, such as absorbed hydrogen in the metal resulting from the pickling operations, which undoubtedly have some effect on the fatigue properties of the metal, may well be reflected in our empirical values. These effects, however, are not sufficiently well understood to warrant any general or detailed conclusions at this time.

ACKNOWLEDGMENT

The authors wish to thank Professors Bradley Stoughton, Gilbert E. Doan and Allison Butts for counsel and encouragement throughout the course of this investigation. This research was supported by funds of the New Jersey Zinc Company Research Fellowship.

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## DISCUSSION

**Written Discussion:** By F. N. Speller, advisory engineer, National Tube Co., Pittsburgh.

This paper gives some useful data which indicates that steel can be pickled, with due attention to the use of proper pickling inhibitors, bath temperature and time of pickling, without reducing its fatigue limit. Several good acid inhibitors are now available on the market besides the ones mentioned by the authors.

Removal of mill scale can also be facilitated by exercising control over the formation of the scale so that the pickling time can be reduced. Mill scale removal is sometimes desirable to put the metal surface in the best condition to receive and hold protective coatings. Where the coated metal is subsequently subjected to fatigue stresses it is desirable that the etching effect be uniform and that sharp notches caused by overpickling be avoided.

**Written Discussion:** By Walter R. Bloxdorf, metallurgist, MacWhyte Co., Kenosha, Wis.

Upon reviewing the various results of this interesting investigation, the writer was particularly impressed by those showing the effect of sulphuric acid concentration on fatigue life in Fig. 11. These results definitely indicate that one engaged in production pickling of this type of steel need not be too greatly concerned over slight or periodic fluctuations in actual acid concentrations of his bath. Indeed, it is striking that such a wide variation in concentration has such an apparently negligible effect upon the fatigue limit of the pickled product. The range in concentrations investigated is much greater than the normal fluctuation which would exist in a given solution in a well regulated pickling room.

From a careful study of the notch conditions shown in Figs. 12a, 12b, and 12c, it appears that when employing similar time and temperature of pickling, the higher concentrations of acid deserve careful consideration on the part of those engaged in pickling material which is to be subsequently cold-worked. The notch condition shown in Fig. 12c would probably result in a cold-worked article having a smoother surface than that produced from pickled material possessing the type of notch shown in Figs. 12a and 12b. It is the writer's belief, also, that the fatigue properties of the cold-worked articles made from material having the notched condition shown in Fig. 12c would be superior.

Fig. 13 showing the effect of pickling bath temperature on fatigue life reveals both interesting and assuring information. It points out quite clearly that the current trend in reducing where possible and permissible the temperature of pickling solutions is proper. Although the results shown cover the work done with an unhibited acid solution, I would assume that similar results would be obtained by using an acid solution properly inhibited.

The writer was somewhat disappointed upon learning that the part of this investigation dealing with what effect ferrous sulphate content of the solution had upon fatigue life yielded rather poor and inconsistent results. This is unfortunate in that, when operating a sulphuric acid solution in production practice, it must be remembered that the ferrous sulphate content is a factor which is forever present and, under certain conditions of pickling, should be under control. It has been our experience that upon pickling patented rope wire steels having a carbon content exceeding 0.65 per cent using an inhibited sulphuric acid solution of about 10 per cent by weight at temperatures of 150 degrees Fahr. or greater, as the ferrous sulphate content increased, the pickled surface of the steel became rougher and the tendency toward pit formation was pronounced. Upon checking the fatigue properties of wire cold drawn from this steel, it was observed that as the pickled surface of the patented material became rougher the fatigue life of the resultant wire became lower. To be sure, the decrease in fatigue life resulting from this roughened surface was slight, but nevertheless it was enough to warrant consideration. Accordingly, we lowered our temperature of pickling to about 120 to 125 degrees Fahr. and found that this change eliminated to a noticeable degree the tendency of the pickling solution to roughen and pit the surface as its ferrous sulphate content increased. The fatigue life of the cold drawn wire was, also, found to be practically constant regardless of the ferrous sulphate content of the bath, provided, of course, that the amount of ferrous sulphate did not exceed a certain figure recognized as a maximum for good pickling practice. Needless to say, at this lower temperature it required a greater time to accomplish a given degree of pickling, but this practice constituted one method for controlling the adverse effect which increasing amounts of ferrous sulphate in our pickling solutions was known to have upon our cold drawn wire.

### Authors' Reply

We appreciate the interest shown in our paper and the discussion submitted by Messrs. Speller and Bloxdorf.

We are in full accord with Mr. Speller's comments and would like, in